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SUBSTITUTION FOR COBALT AND CHROMIUM IN THE AIRCRAFT GAS TURBINE ENGINE

by

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Industry, Technology, and Employment Program

September 1983

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J952-84

SUBSTITUTION FOR COBALT AND CHROMIUM

IN THE AIRCRAFT GAS TURBINE ENGINE

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INTRODUCTION

Gas turbine engines exist in a variety of forms and are used in many diverse applications to develop power. This power is produced either as jet thrust power (i.e. as propulsive power delivered by a high velocity hot gas jet exhaust), as shaft power (i.e., torque in a rotating shaft), or as a combination of the two. In aviation the basic forms of the gas turbine are the turbojet, the turbofan, the turboprop, and the turboshaft engine. The turbojet and turbofan produce jet thrust power and are used to propel jet airplanes and some missiles (Figs. 1 and 2)--these engines are referred to collectively as jet engines. The turboprop engine (Fig. 3) develops shaft power to drive a propeller, and only a very small amount of thrust power is derived from the hot gas jet exhaust. The turboshaft engine (Fig. 4) develops shaft power and virtually no thrust is derived from its exhaust; these engines drive helicopters and are also used on board aircraft as auxiliary power units (electrical and hydraulic power generation).

Use of the gas turbine engine is not limited to aviation. The turboshaft engine has many other uses:

1. To drive electric power generators, sometimes in conjunction with a steam turbine.
2. To drive centrifugal compressors that pump gas along gas pipelines.
3. In the petroleum industry, to drive gas compressors that maintain well pressure, and to drive refinery air and gas compressors.
4. To drive air compressors for the operation of blast furnaces in the steel industry.
5. To provide propeller shaft power for ships; on board ships they are also used as auxiliary power units to drive electric power generators, fire pump systems, bilge pump systems, and compressors for pneumatic power.
6. To provide motive power for automobiles and trains.

While the gas turbine is often thought of as a fairly recent technical development, being usually associated with the advent of jet airplanes around 1940, the basic concept is considerably older (1,2).

Gas turbines for power use were suggested in the 19th century and industrial sets were in production in Switzerland in the mid-1930s. In fact, in the patent literature gas turbines appear at intervals back to a British proposal of 1791.

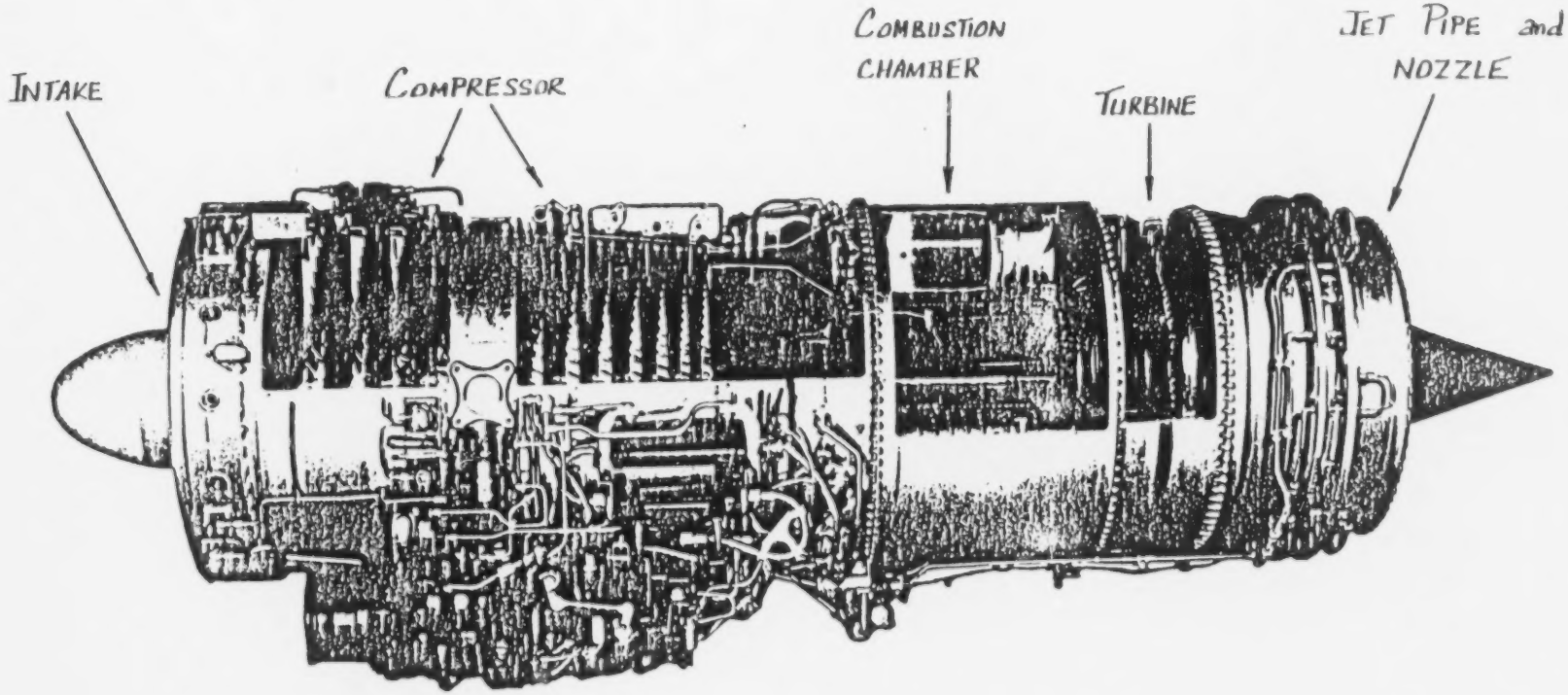


FIG. 1 Pratt & Whitney J52 Turbojet

Source: United Technologies

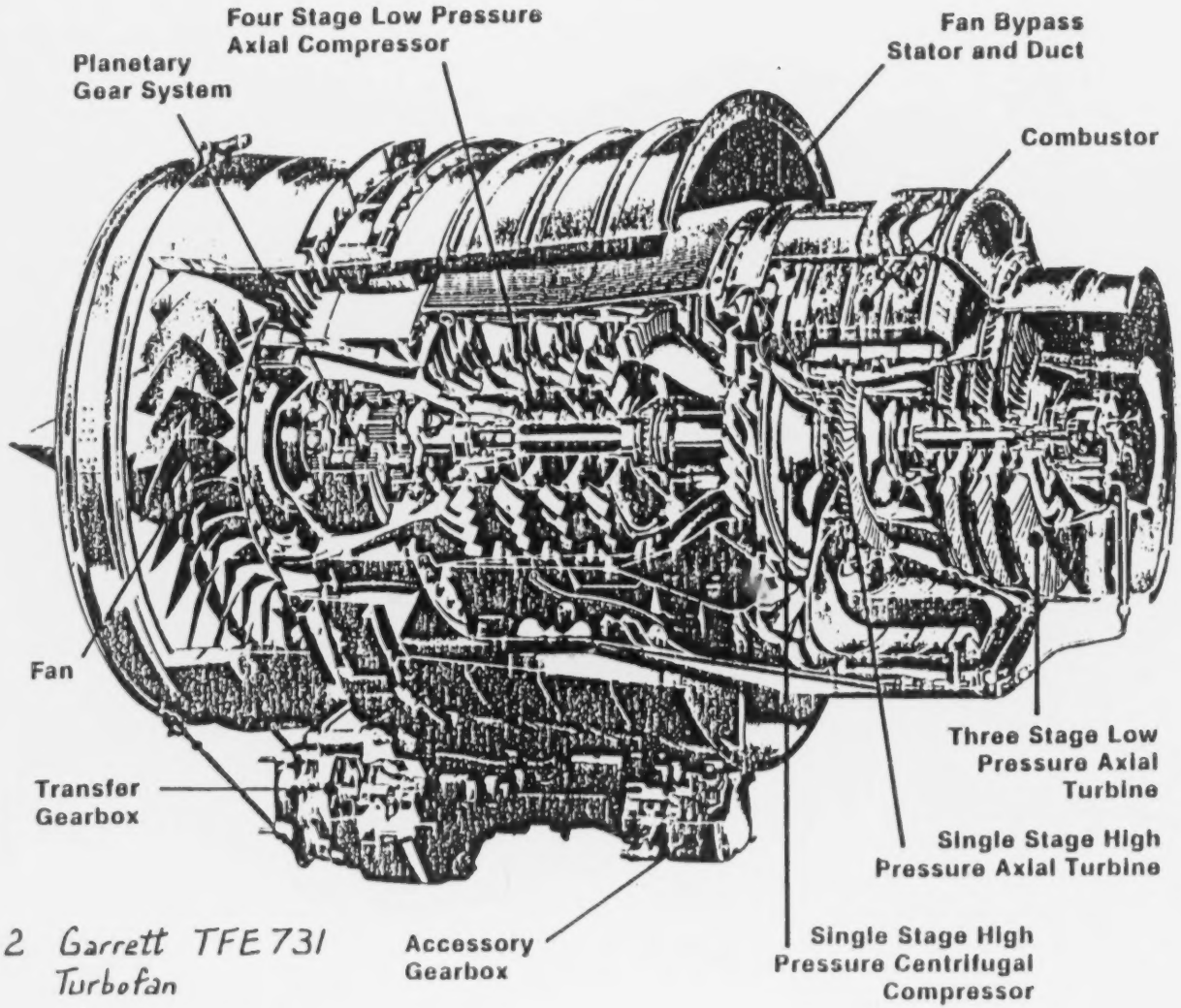


FIG. 2 Garrett TFE 731 Turbofan

Source: Garrett Turbine Engine Company

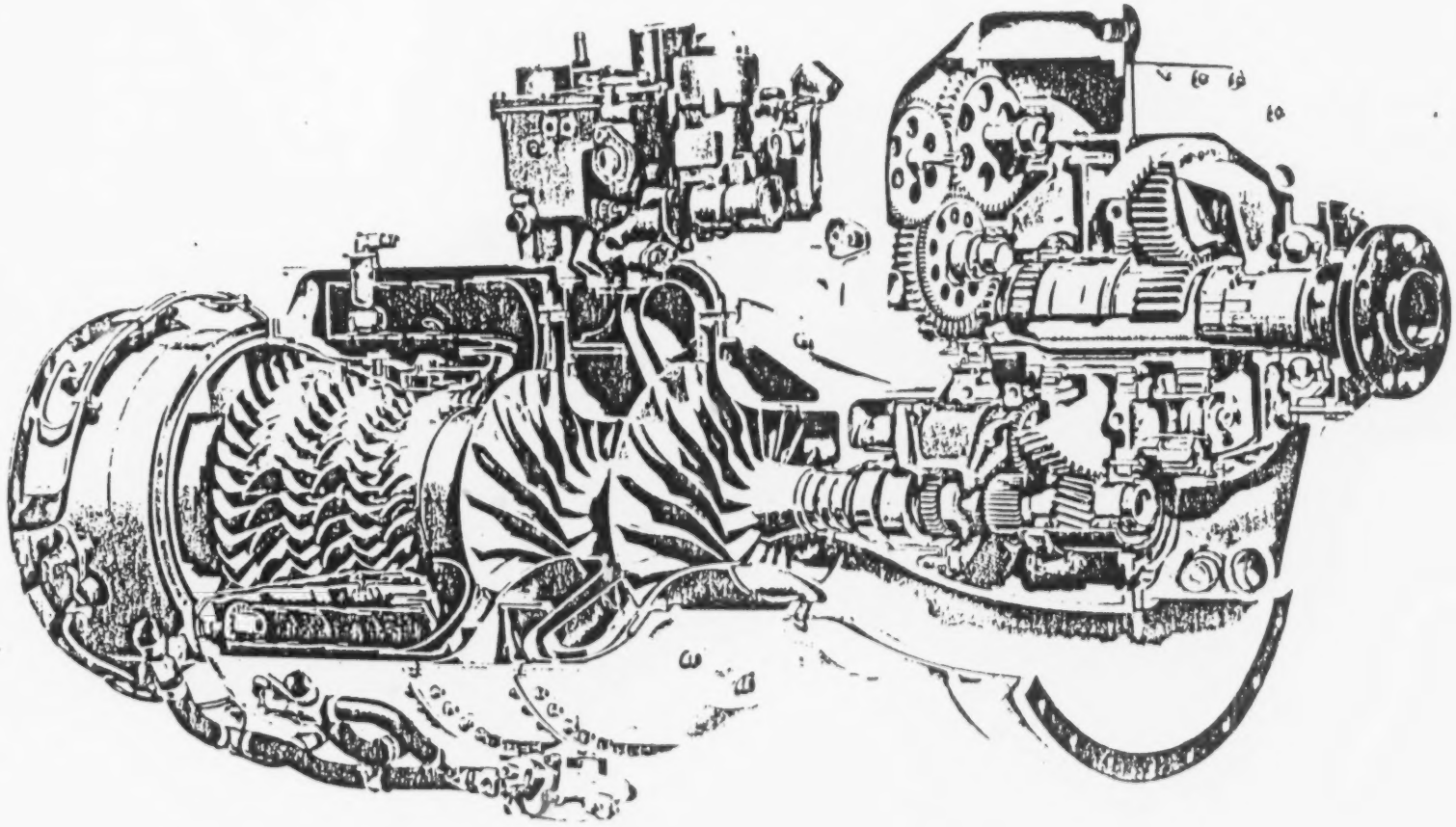
> In the field of aircraft propulsion, various proposals appeared from about 1910 for the use of some form of turbine power plant instead of a piston-engine. Serious engineering calculations for gas-turbine aircraft propulsion began at the Royal Aircraft Establishment at Farnborough in the mid-1920s, and H.H. Constant and A.A. Griffith...published papers on the characteristics of turboprops, in which the turbine would be used to drive a conventional propeller. But it was the unknown cadet at Cranwell, Frank Whittle, who proposed [in 1928] to combine the use of a gas-turbine with elimination of the propeller to produce the turbojet (2).

Whittle patented his idea for the turbojet in January 1930. His design was first tested on the ground on 12 April 1937, and after considerable development effort first flew on 15 May 1941.

In parallel with these British efforts, the Germans separately developed the aircraft turbojet. Despite the fact that Whittle ran his engine first, the Germans flew theirs first--on 27 August 1939.

In the spring of 1941, the year Whittle's engine first flew, the General Electric Company in the United States was asked by General H.H. Arnold (U.S. Army Air Force Chief of Staff) to produce a jet engine for a fighter airplane. This first U.S. jet engine (the General Electric I-A) was based on the Whittle turbojet and first flew on 1 October 1942. While the I-A was being developed, the National Advisory Committee for Aeronautics (NACA) let a contract to the Westinghouse Corporation in late 1941 for the development of an American engine. Westinghouse was not told of the existence of the Whittle engine. Within a short time other U.S. companies entered the gas turbine engine busines. Present U.S. manufacturers are General Electric, Pratt & Whitney Aircraft, Garrett Turbine Engine Company, the Allison Gas Turbine Operations of General Motors, Teledyne Continental, Teledyne CAE, Avco Lycoming, Solar, and the Williams Research Company.

It is clearly desirable that the gas turbine engine deliver maximum power for a given size in whatever application it is used. Further, for aircraft propulsion it is a major design goal to maximize the thrust/weight ratio and fuel-efficiency of the engine in order to maximize airplane performance. The power and fuel-efficiency of the gas turbine engine increase with the temperature achieved at the inlet to the turbine, and with the efficiency of the various engine components. Achievement of the maximum possible turbine



TPE331 TURBOPROP ENGINE
GARRETT TURBINE ENGINE COMPANY

MS 7330

FIG. 3

inlet temperature (TIT) is therefore always very important; for the aircraft engine this must be done with the lightest possible structure.

The dual requirement of high temperature and light weight places severe demands upon the materials from which certain components of the engine are made. Many parts are subjected not only to high temperatures but also to high mechanical and thermal stresses; both temperature and stress are imposed in cycles over which they vary considerably causing repeated expansion, contraction, and deformation of the parts; some parts are exposed to severe corrosive environments such as oxidation and hot-corrosion. The parts must be capable of operating reliably in the midst of this nightmare for a minimum specified period of time, termed the required life of the part. The required life of various components has increased significantly over the years. The performance of the gas turbine engine--as measured by its maximum thrust or shaft power, thrust or power/weight ratio, thrust or power/frontal area ratio, fuel consumption, component life, and maintenance requirements--is very dependent upon the materials used in its construction.

When Whittle proposed the concept of the turbojet in 1928 there were no materials suitable for the hottest of its hot-section parts (2). Ever since that time, jet engine development has been largely dependent upon the development of such materials, and steady advances have been made in the maximum operating temperature of these "high temperature materials." The search for metals in this category led to the development of the **superalloys** which obtain their remarkable high temperature properties from the alloying of about a dozen particular elements, chief amongst which are nickel, cobalt, chromium, aluminum, and titanium. The operating conditions for which a given superalloy is suitable vary with its composition and manufacturing history so that it is useful for the fabrication of only particular components of the engine; there are thus many different superalloys in use. As the superalloys have grown more exotic their costs have escalated--the development of a new alloy now takes about 3-10 years and costs from $2-10 million depending on the application. The cost of these materials enhances the desirability of long component life.

The metals that are presently considered most strategic by the United States aircraft gas turbine industry are cobalt (Co), chromium (Cr), columbium

(Cb), and tantalum (Ta) (3). U.S. consumption of these metals during 1979 is shown in Table 1; the airplane gas turbine industry consumes the majority of the superalloys.

Each of the elements Co, Cr, Cb, and Ta plays a different role in enhancing the properties of a given superalloy and one or more of them may indeed be absent. In some current superalloys it is possible to substantially reduce the amount of one or more of these elements without significantly reducing performance. But in many cases such reduction would necessitate a reduction in engine performance--operating temperature, component stress, or component life. In fact, as new generation engines are being developed with higher temperatures, some alloys with no Co are being replaced with Co containing alloys. The aircraft gas turbine engine manufacturers consider it impossible to remove the four elements Co, Cr, Cb, and Ta entirely from all their superalloys, for the performance losses would be intolerable.

There are several ways other than the reduction of Co, Cr, Cb, and Ta levels in current superalloys by means of which the hot-section consumption of these strategic elements may be reduced. These include the use of alternate superalloys, the wider use of coatings, the possible future substitution of various metallic and non-metallic high temperature structural materials presently under development, and substitution by design.

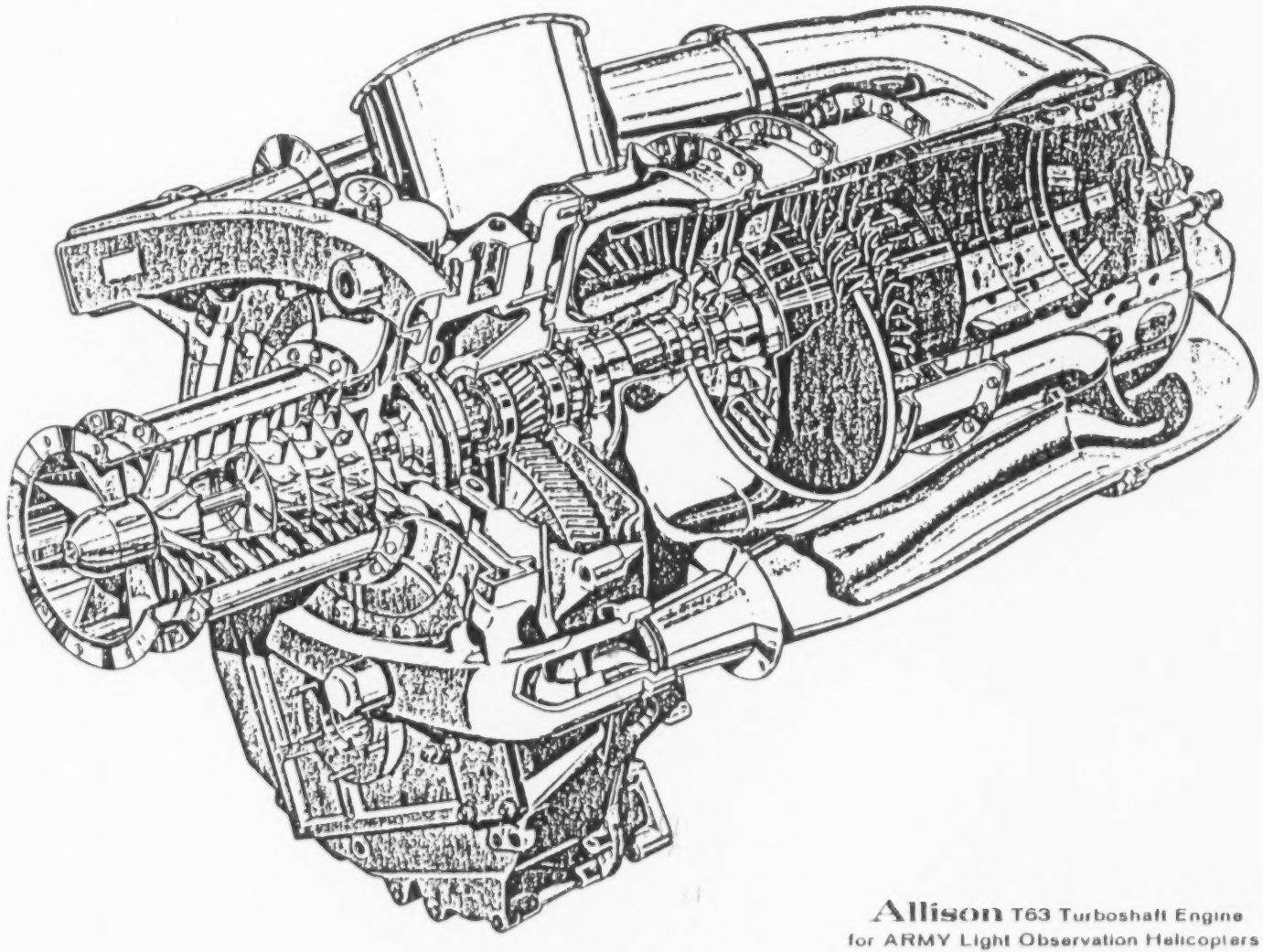
However, it appears that until non-metallic materials (such as ceramics and carbon-carbon composites) are sufficiently well developed that they can be used to construct hot-section parts, aircraft gas turbine engines will require the use of Co, Cr, Cb, and Ta. This development is presently seen by the National Aeronautics and Space Administration (NASA) as occurring beyond the year 2000 for some rotating hot-section parts in man-rated engines. The actual production dates for engines containing significant quantities of non-metals in their hot-sections may be well beyond the year 2000 since the next generation of engines on the drawing boards now will not see production until the mid-1990s. Therefore in discussing the issue of strategic material substitution in the aircraft gas turbine engine, we must address two time periods: the next 20 years (approximately) during which metals (mostly superalloys) will be used for the majority of hot-section parts; and the subsequent era when non-metallic hot-section parts may be used. During the

Table 1

United States Consumption of Strategic Materials (1979)

Metal	Total Consumption in All Industries (millions of lb)	Consumption in Superalloys	
		%	(millions of lb)
Co	20.3	30	6.09
Cb	6.3	28	1.76
Ta	1.7	5	0.085
Cr	1000	3.4	34.0

Source: Ref. 3



Allison T63 Turboshaft Engine
for ARMY Light Observation Helicopters

FIG. 4

first period there is limited potential for reducing strategic material content of such parts, and during the second it is possible that metals currently regarded as strategic may not be required in the fabrication of those parts at all.

The following discussion addresses these matters in some detail. It is restricted to airplane gas turbine engines, focusing on the use of and substitution for Co and Cr in the hot-section of those engines. It first reviews the present configuration and operation of the aircraft gas turbine engine. This is followed by a discussion of the materials property requirements of hot-section parts. Materials used in engine construction are then discussed; this includes a brief history of materials usage for the major engine components, but concentrates on present day and possible future materials for several major hot-section parts. The roles that Co and Cr play in enhancing the properties of hot-section materials are then identified. Finally, the possibilities for materials substitution to reduce the use of Co and Cr in various hot-section parts are discussed.

DESCRIPTION OF THE AIRCRAFT GAS TURBINE ENGINE

The four classes of gas turbine engines--turbojet, turbofan, turboprop, and turboshaft--are basically similar in configuration. The turbojet is described below and major variations in configuration of the other classes are noted.

The turbojet may be divided into six main regions of gas flow: intake, compressor, combustion chamber, turbine, jet pipe, and nozzle (Fig. 1). The hot-section comprises the last four regions. Atmospheric air enters the intake at approximately the flight speed of the aircraft; if the aircraft flies at supersonic speed the entering air is slowed in the intake (and consequently its pressure and temperature are raised) by the intake geometry so that it reaches the compressor at subsonic speed.

There are two types of compressor--centrifugal and axial. Centrifugal compressors are capable of compressing the air through a ratio of about 8:1 and were used on all the earliest successful turbojets; they are used on small present day gas turbines (Figs. 2 and 3). For high performance in large

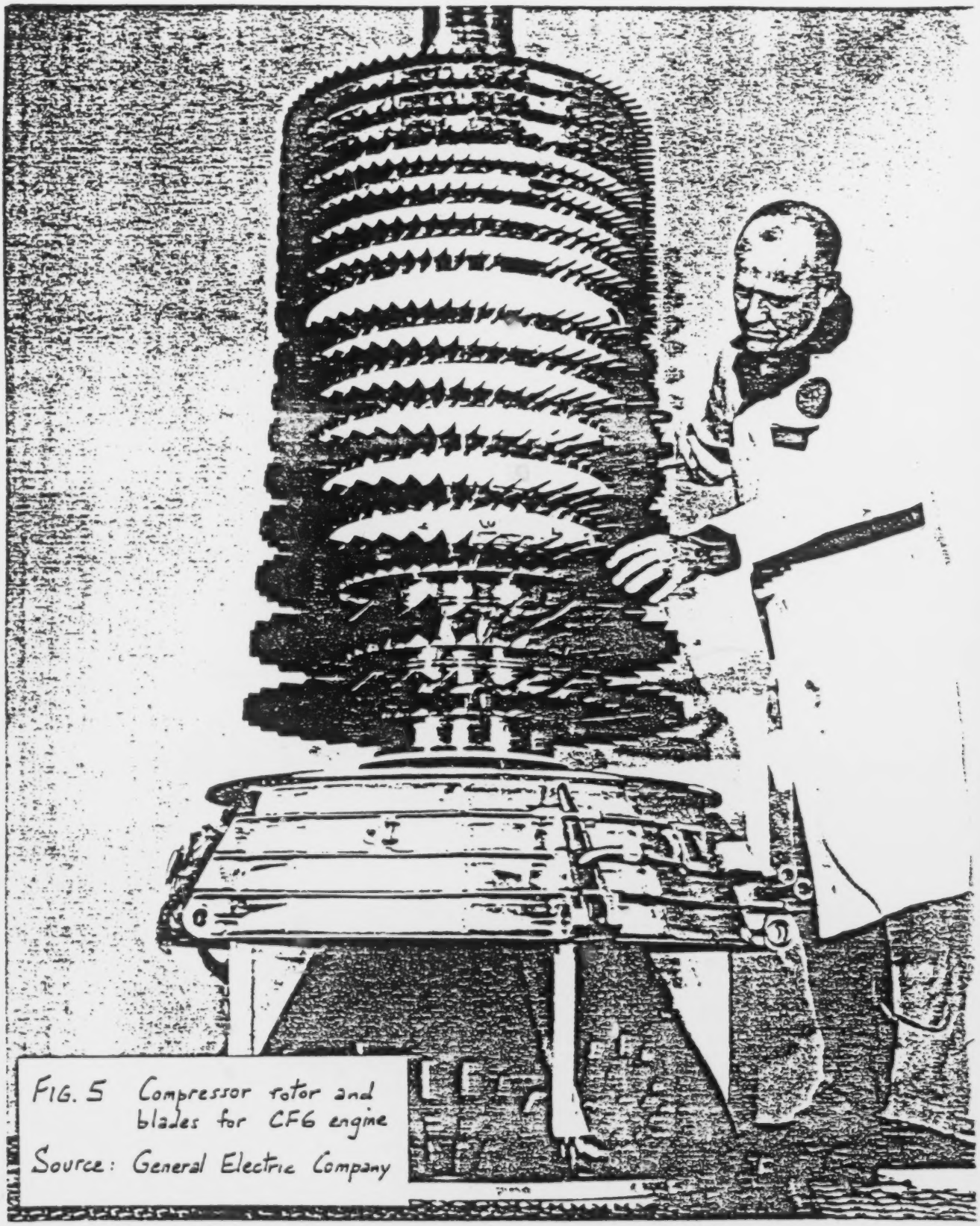


FIG. 5 Compressor rotor and blades for CF6 engine

Source: General Electric Company

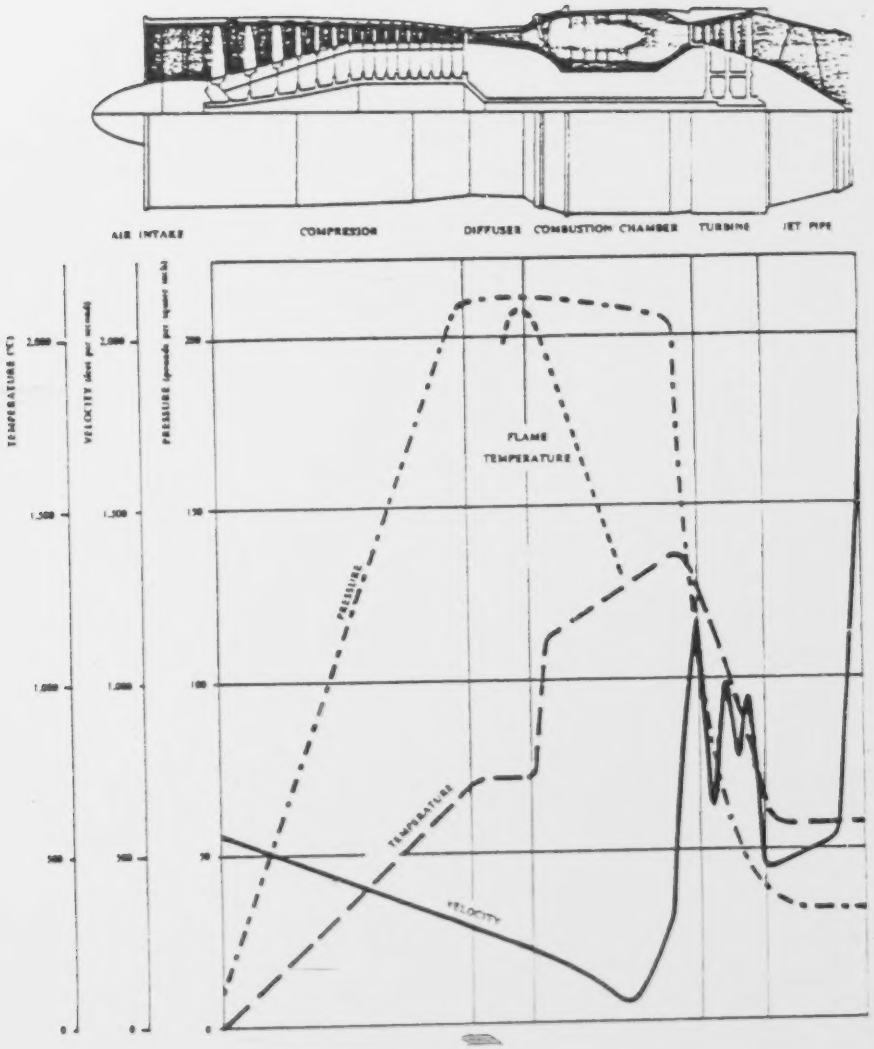


FIG. 6 TYPICAL OPERATING CONDITIONS INSIDE AN AIRCRAFT TURBOJET

SOURCE: REF. 2

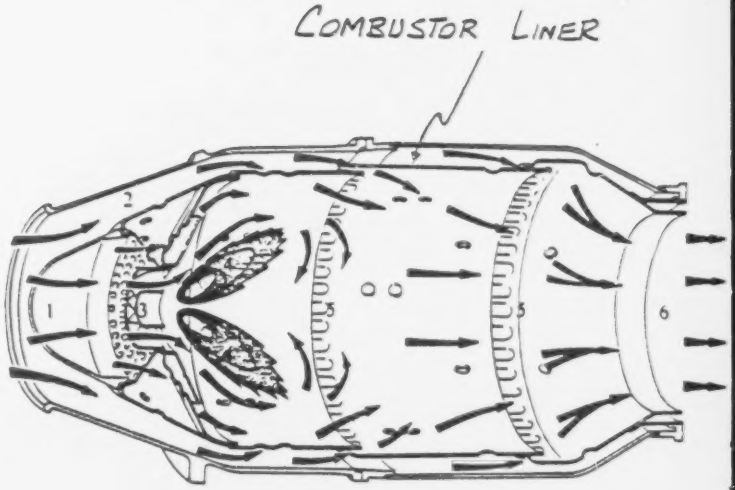
engines an axial compressor is required (Fig. 1). These devices compress the air through a series of blades mounted in parallel rings (stages) on the surface of a rotating cylinder (compressor rotor)--see Fig. 5. Each blade is a small airfoil (wing) projecting radially from the surface of the rotor. The air flows generally parallel to the rotor axis and may be compressed through a pressure ratio around 30:1 by the time it leaves the last stage. Between each stage of blades is a ring of stationary blades (airfoils) called stator blades or vanes which straighten the flow and introduce it to the next compressor stage at the correct angle. As the air passes through the compressor, its temperature rises progressively (this is the result of the work done by the compressor in compressing the air). By the time it exits the last compressor stage its temperature may have risen as much as 700C or 1300F.

Typical variations of gas velocity, temperature, and pressure in a stationary turbojet with an axial compressor and axial turbine are shown in Fig. 6.

As the air leaves the compressor it enters the **combustion chamber**--sometimes called the **combustor** or **burner**--where it is mixed with fuel and burned; a typical flame temperature is 2000C or 3600F. Some engines use can-type combustors (Fig. 7) arranged in a ring about the engine centerline behind the compressor, with the shaft that drives the compressor passing along the centerline. Other engines use a single annular chamber encircling the engine shaft (Fig. 7). Only some of the air entering the combustor is actually burned. Much of it is directed around the flame zone and is mixed with the burned gas downstream in the combustor thereby diluting it. Some of the diluting air enters through the side walls of the liner and cools the liner enroute. Within the combustor the gas pressure is virtually constant at the value it attained at the compressor exist, but the velocity and temperature vary considerably. The diluted gas at the combustor exit has a typical mean temperature of 1650C or 3000F.

Upon exiting the combustor the hot gas enters the **turbine** which is usually an axial flow turbine (Fig. 8) although radial turbines are used in some small engines. An axial turbine comprises a number of stages of turbine blades (airfoils), each stage mounted around the rim of a circular disc; each blade is retained on the disc by a dovetail seration called a firtree (Figs. 9

Air from the compressor enters through swirl vanes, and a perforated flare to support primary combustion at an air/fuel ratio of about 5:1. The bulk of the air (over 72 per cent of the total) then enters through apertures farther along the flame tube, to dilute the flame to an air/fuel ratio of from 15:1 to 140:1 and to yield a steady stream of hot gas at a uniform temperature of some 900° to 1,400°C which can be used as the medium both to drive the turbine and, in a jet engine, to provide the propulsive thrust.



1 Primary air snout
2 Secondary airflow (about 80 per cent of total)
3 Fuel injector here
4 Flame cone from finely divided fuel spray
5 Dilution and cooling air
6 Even temperature hot gas to drive turbine

a) TYPICAL CAN-TYPE COMBUSTION CHAMBER
SOURCE: Ref. 2

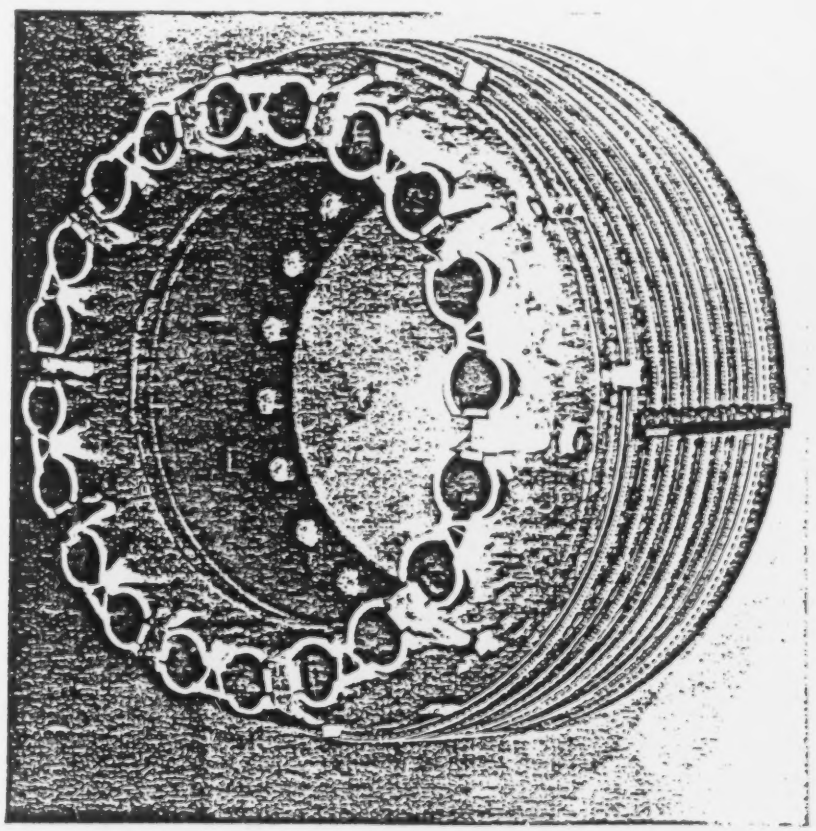


FIG. 7

b) TYPICAL ANNULAR COMBUSTION CHAMBER
Source: United Technologies
Pratt & Whitney Aircraft

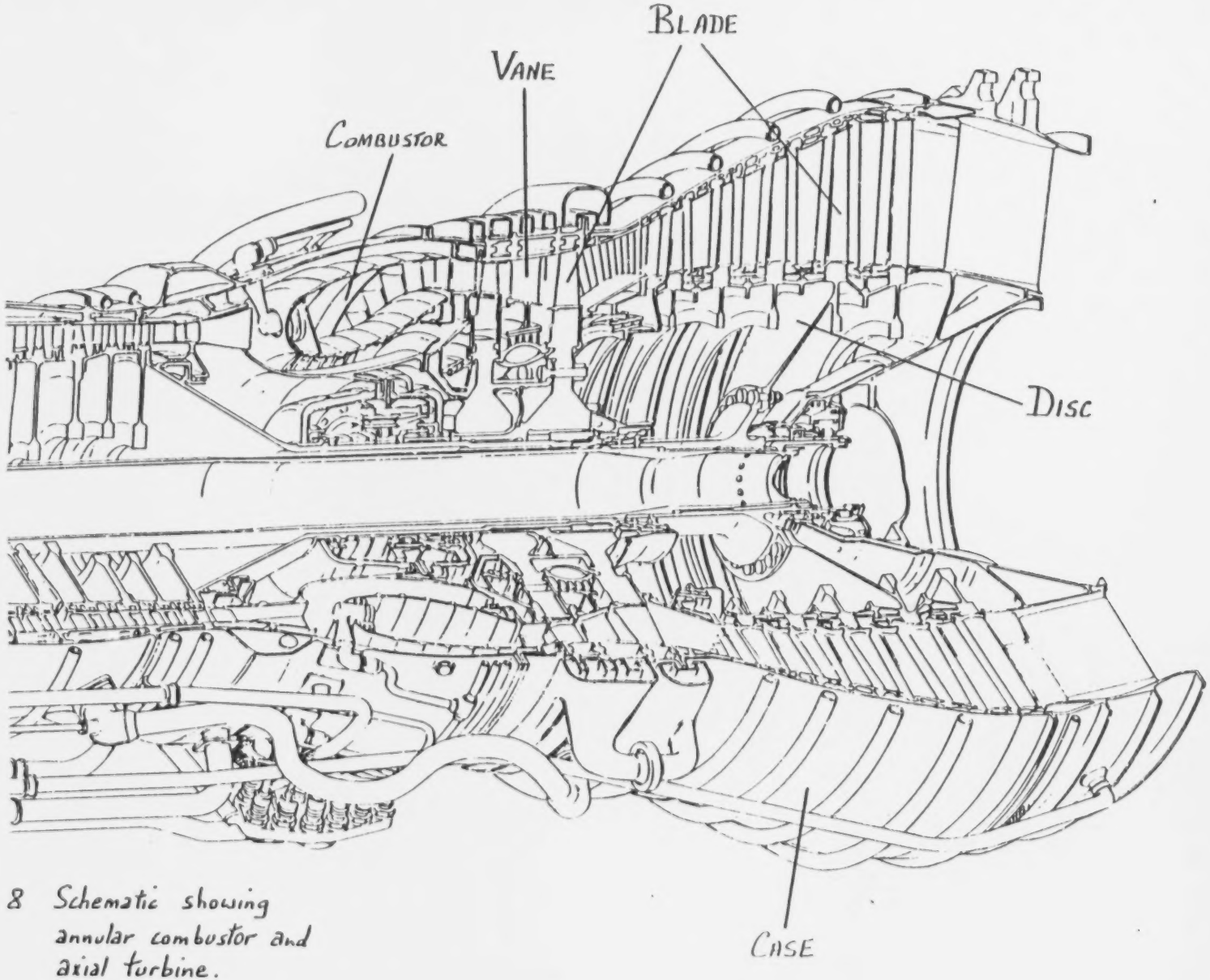


FIG. 8 Schematic showing annular combustor and axial turbine.

Source: United Technologies Pratt & Whitney Aircraft

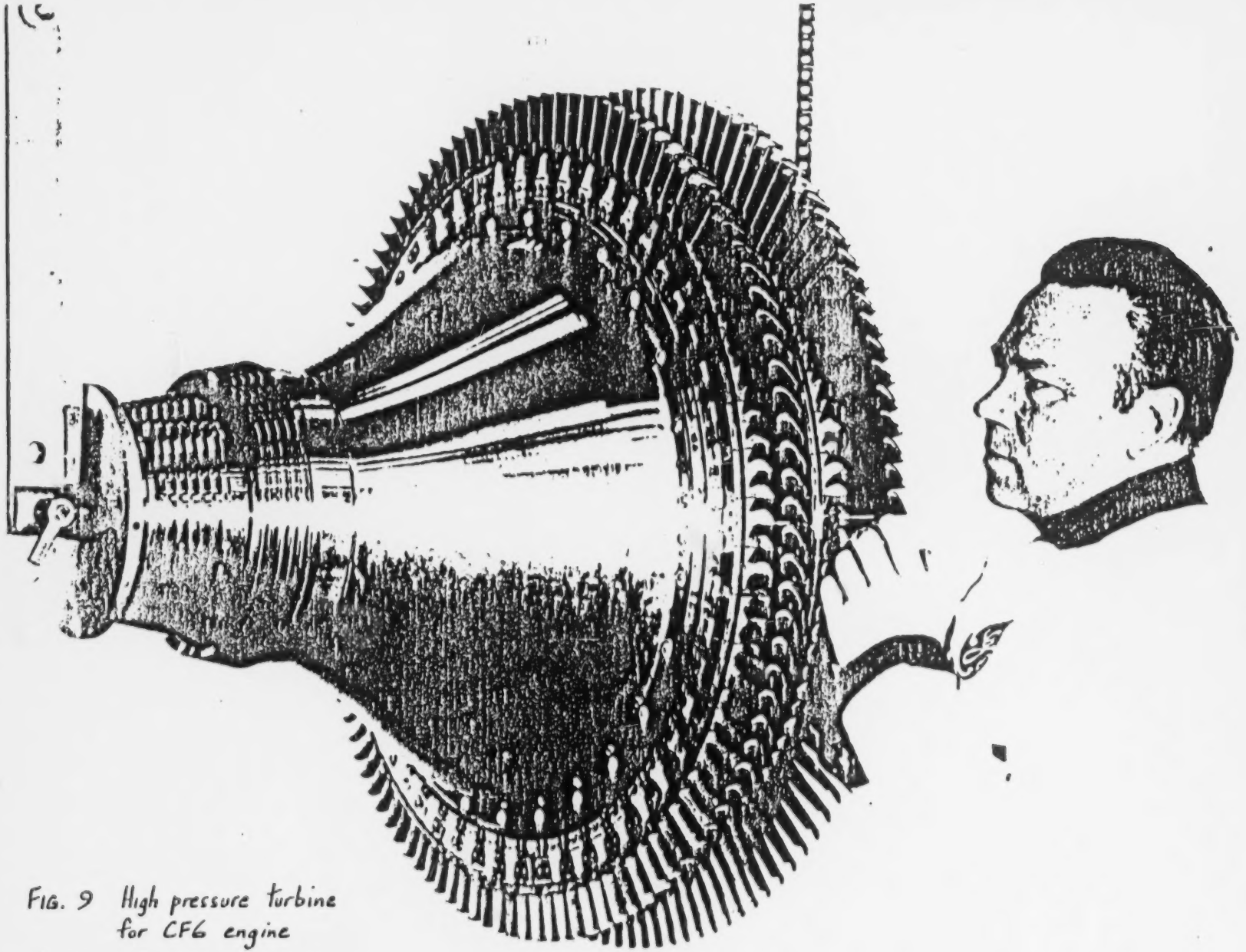


FIG. 9 High pressure turbine for CF6 engine

Source: General Electric Company

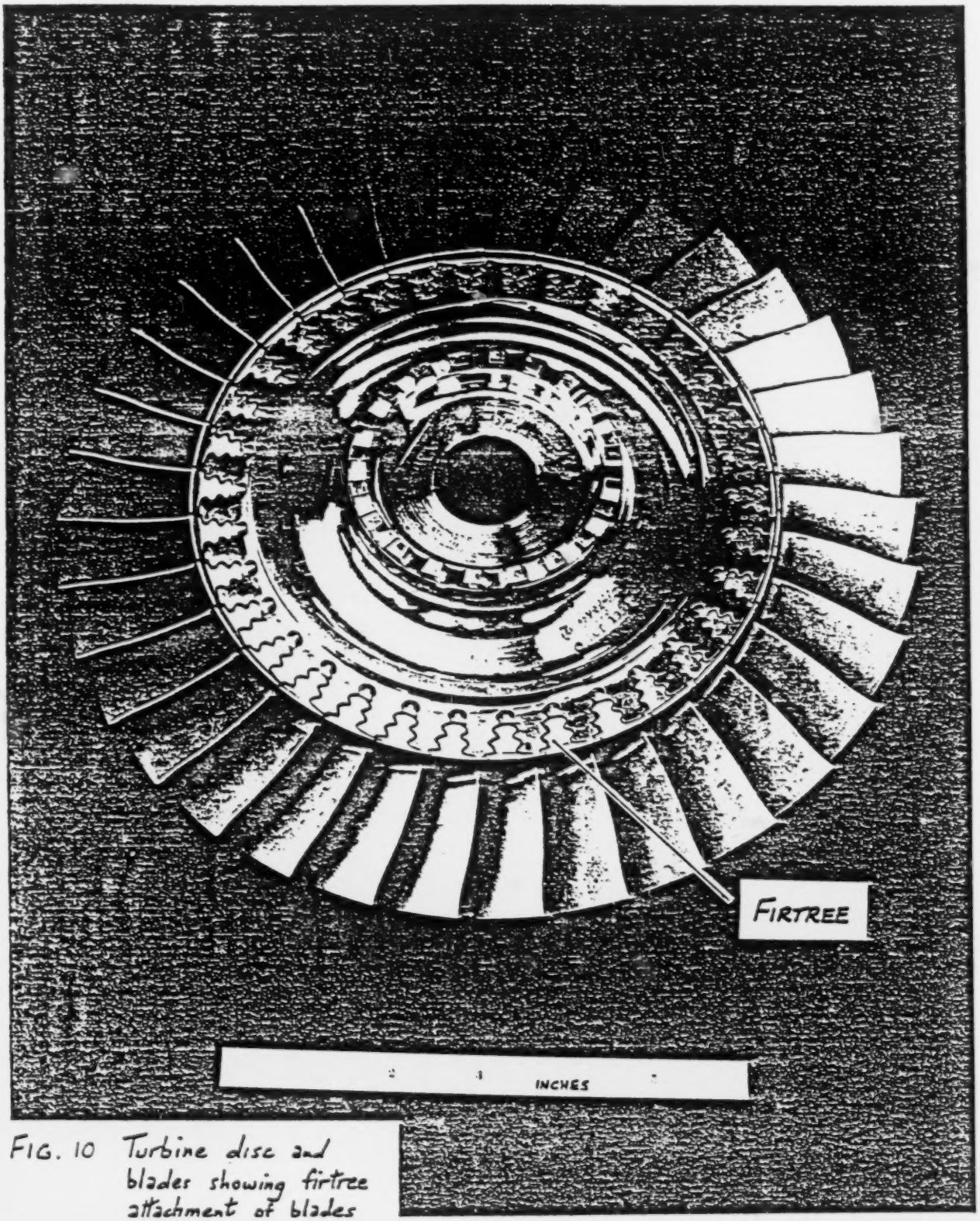


FIG. 10 Turbine disc and blades showing firtree attachment of blades

Source: Garrett Turbine Engine Company

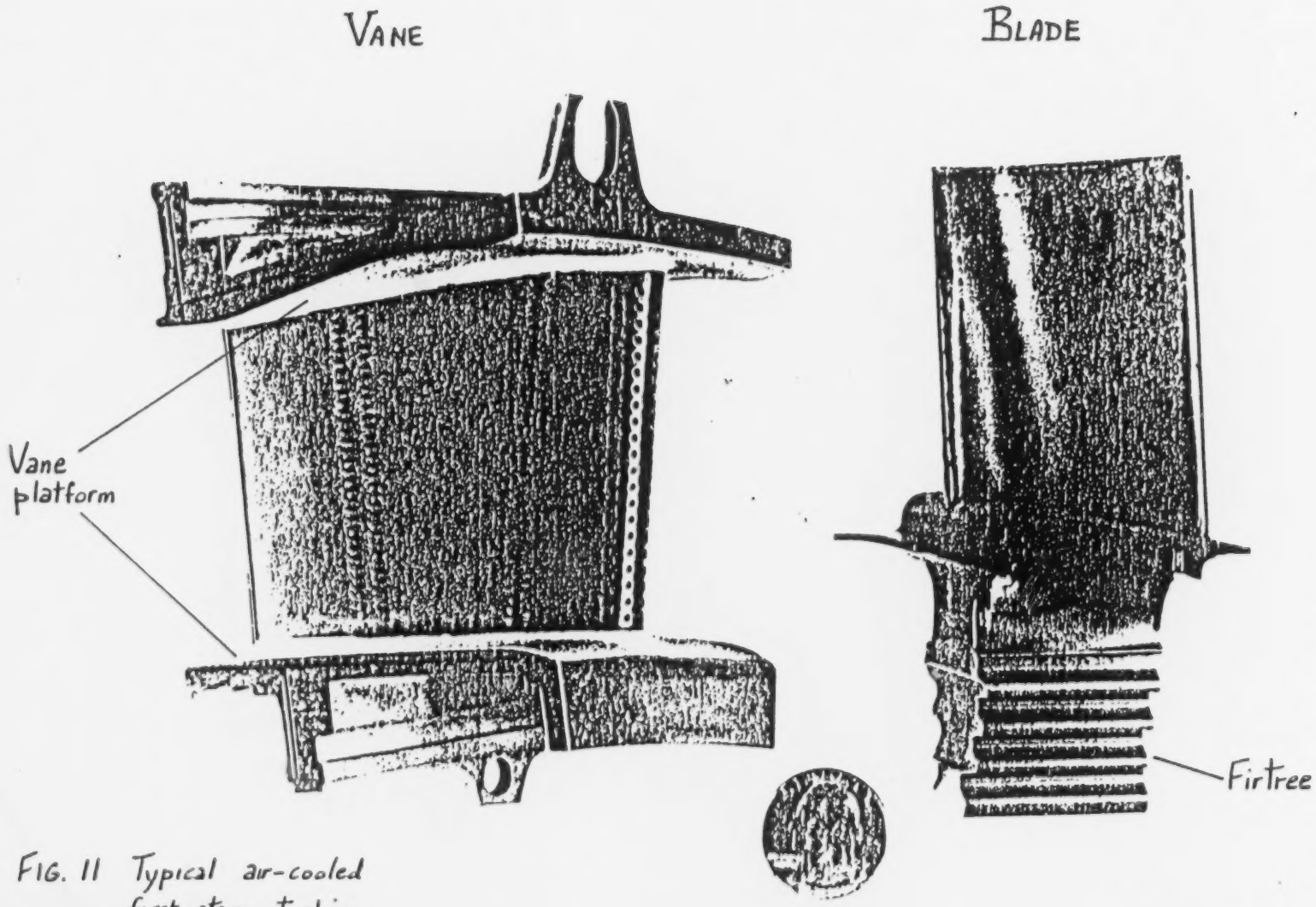


FIG. 11 Typical air-cooled first stage turbine vane and blade

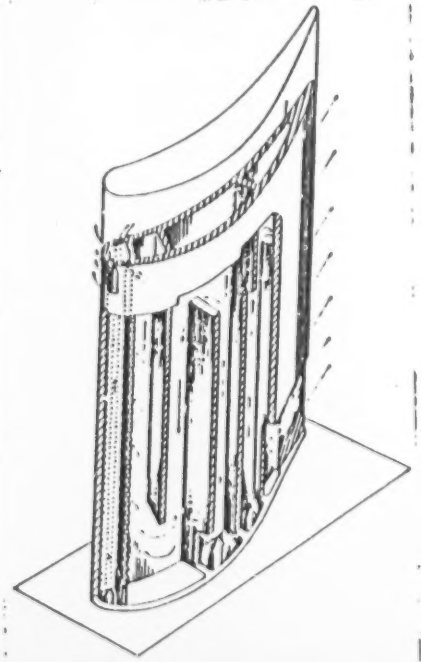
Source: United Technologies

and 10). Each stage of blades is separated from the next by a ring of stationary blades called vanes (Fig. 8). A typical turbine blade and vane are shown in Fig. 11; because of the high temperatures at which these components must operate they are often cooled by passing compressor bleed air through complex internal cooling passages and out through pinholes in their exterior surfaces. A typical turbine blade cooling arrangement is shown in Fig. 12 (4). Turbine discs are often cooled by blowing compressor bleed air on their foreward and aft faces. The turbine has the inverse function of the compressor--it removes energy from the hot gas thereby decreasing its temperature, velocity and pressure. In the turbojet this removed energy is used to drive the compressor by means of a rotating shaft rigidly attached to the turbine and compressor, and represents only a small fraction of the total energy of the hot gas. In the turbofan, much energy remains in the hot gas after the turbine, that removed by the turbine being used to drive the compressor and fan. The fan is essentially a propeller with many blades in a duct. In the turboprop and turboshaft engines, the maximum possible energy is removed by the turbine; this energy is used to drive the compressor and propeller in the turboprop, while in the turboshaft engine it is used to drive the compressor and whatever device is attached to the engine output shaft.

In some engines all of the turbine discs may be connected together into a single unit and this used to drive a single unit (but multistage) compressor plus fan, propeller, or output shaft. Such an engine is called a single-spool engine. Many engines have two spools, the turbine discs being connected together in two groups, and the compressor stages being mounted on two separate rotors. Passing through the engine from front to rear, the first group of compressor stages is called the low pressure compressor, the second the high; the first group of turbine stages is called the high pressure turbine, the second the low. The high pressure turbine is then used to drive the high pressure compressor through one shaft, and the low pressure turbine drives the low pressure compressor (plus the propeller, fan or output shaft) through a second shaft coaxial with the first (Fig. 13). In a three-spool engine a third turbine may drive the fan separately. Multi-spool arrangements permit more efficient operation and higher performance.

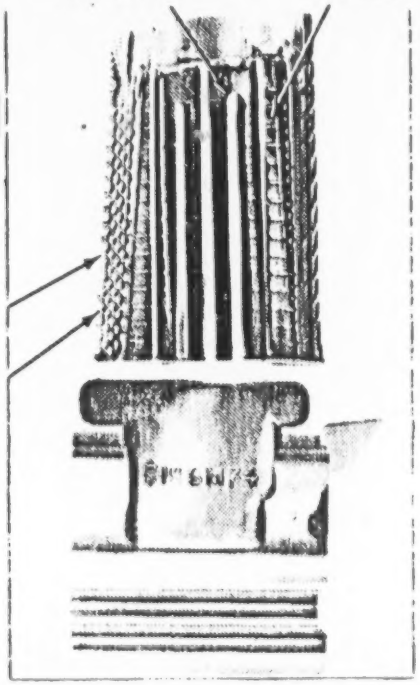
Upon leaving the turbine in a turbojet, turbofan, or turboprop engine the hot gas enters the jet pipe and then the nozzle from which it is exhausted to

Turbine Blade With Complex Internal Cooling System

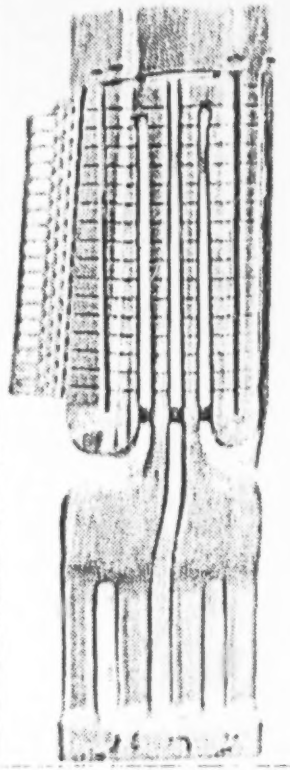


Pin Fins

Trailing Edge Cooling Slots



Partially Machined Casting With Convex Airfoil Skin Removed to Show Internal Passages



Ceramic Core Produces Internal Features During the Casting Process

FIG. 12 Film and convection-cooled blade

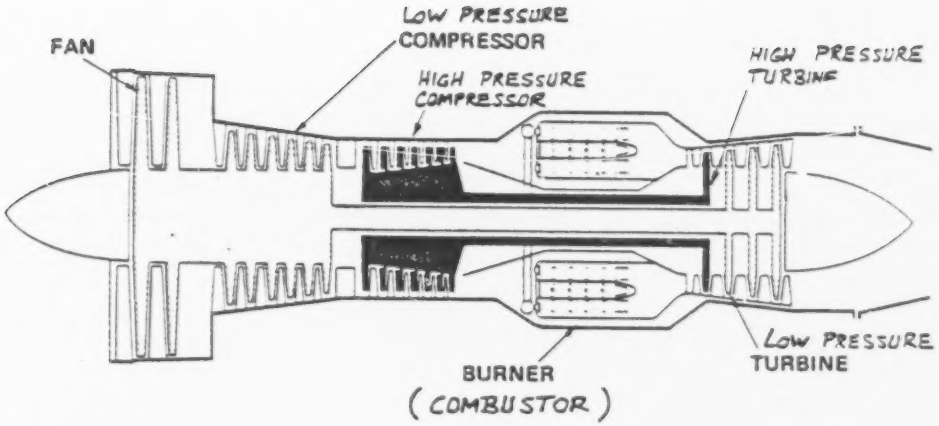


FIG. 13 TWO-SPOOL TURBOFAN ENGINE WITH AXIAL FLOW COMPRESSOR AND AXIAL FLOW TURBINE

SOURCE: ADAPTED BY OTA FROM REF. 8

atmosphere. In a fighter airplane jet engine, the jet pipe may incorporate an afterburner wherein more fuel is injected and burned in order to further raise the temperature and pressure of the hot gas. In the nozzle as much as possible of the heat and pressure energy in the hot gas is converted to kinetic energy (high velocity) in order to maximize engine thrust. The turboshaft engine does not employ a nozzle. It may employ a regenerator to remove heat energy from the turbine exhaust--this energy may then be used to heat gases upstream in the engine thereby increasing engine efficiency and shaft power.

In order to retain the pressures developed in the engine from the fan to the nozzle, the gas flow passages are surrounded by a pressure-bearing tube called the engine case. The case comprises various sections, each constructed to withstand the local conditions of temperature and pressure. These sections are fabricated as frameworks of load-bearing members to which are attached rolled sheet or cast or forged shells that contain the high pressure gas (Fig. 8).

MATERIALS REQUIREMENTS OF HOT-SECTION PARTS

The hot-section of the engine comprises the combustor, turbine, jet pipe, and nozzle. The operating conditions of the various parts therein vary considerably and consequently so do the property requirements of the materials from which they are made.

Apart from being exposed to high temperatures, the parts are variously exposed to a number of different forces, both cyclic and steady, including: gas forces; inertia forces due to rotation and vibration; gyroscopic forces arising from a combination of rotation and aircraft maneuvering; impact of foreign objects passing through the engine; forces due to constrained thermal expansion and contraction; the rubbing together of adjacent interfering surfaces; and internal forces due to temperature gradients within parts, especially during engine starting and shutdown. In addition, the parts are exposed to erosion by foreign particles, oxidation, aqueous corrosion, and hot-corrosion due to combustion products and chemicals such as salt in the intake airstream. This complex environment demands a range of material properties which are described below. The importance of each of these

properties varies with the part concerned; and since there is no single known material that fully demonstrates all of the properties desired for some parts, the choice of material often involves compromise.

Description of Materials Properties

The following properties are important for hot-section parts made of a metal.

1. **Strength.** The static stress, applied at a given temperature, necessary to permanently deform a specimen upon release of the stress (yield strength) or to break a specimen (ultimate strength). The applied stress may be either a tensile, compressive, or shear stress, and the material strength varies accordingly.
2. **Ductility.** The ability of a specimen to be permanently deformed without fracture by the application of load, usually tensile. Ductility is measured either as a fractional length change or a cross-sectional area change. Ductility is important during fabrication and service.
3. **Creep Resistance.** When a specimen is loaded in tension, compression, or shear at a sufficiently high temperature it will exhibit a very slow permanent deformation. This deformation is called creep. Creep rate generally increases with load and temperature.
4. **Stress-Rupture Life.** When a specimen is loaded at a sufficiently high temperature it will exhibit creep and will eventually fracture after a time interval called the stress-rupture life. This life decreases with increases in temperature and stress.
5. **Fracture Toughness.** The resistance of the material to the propogation of a crack when loaded with increasing load. High fracture toughness is required for a material to survive impact with a foreign object.
6. **Fatigue strength.** When a specimen is subjected to an alternating stress, small cracks will develop in the specimen. These microcracks grow and combine to form one or more cracks which propogate through the specimen and lead to its eventual failure after a certain number of stress cycles. The number of cycles to failure increases as the mean stress decreases, and as the maximum stress deviation from the mean decreases. The fatigue strength of the material is the maximum applied tensile stress in the cycle (of given mean stress) that results in failure of the

specimen in a given number of cycles. The number of cycles to failure is the fatigue life. The fatigue strength and life are dependent on the material, the stress cycle applied, the geometry and surface condition of the specimen, the temperature, and sometimes the oxidation and corrosion state of the material. There is generally a complex interaction between creep and fatigue at high temperature.

When the cyclic stresses in the material are small (such as those due to cyclic aerodynamic loads on a turbine blade) then fatigue failure will occur only after a very high number of cycles. This failure mode is called **high-cycle fatigue** (HCF). In general, a HCF strength of a material increases with its ultimate tensile strength.

When the cyclic stresses are very large (such as those due to centrifugal load in a turbine blade--the load cycle in this case might be engine start to engine shutdown) then fatigue failure may occur after a small number of cycles. This failure mode is called **low-cycle fatigue** (LCF). The stresses causing LCF may be mechanical (as above) or thermal in origin. When thermal, the LCF is called **thermal fatigue.** The thermal stresses may be due to internal temperature gradients, or to external forces arising from expansion of the part--the thermal fatigue cycle is usually engine start to engine shutdown. A severe thermal fatigue problem may result in failure during a few thermal cycles, perhaps even one; such failures have led to the use of the term **thermal shock** as a synonym for thermal fatigue. Thermal fatigue strength generally increases with ductility, toughness, and thermal conductivity, and with decreasing elastic modulus.

7. **Erosion Resistance.** Small particles in the high velocity gas stream tend to erode parts by abrasive action. These particles arise primarily from poor fuel combustion, but also include sand, dust, and scale deposits dislodged from the combustion chamber. A hard and tough material surface at operating temperature is required to provide erosion resistance.

8. **Wear Resistance.** Two adjacent surfaces that slide or roll over each other will exhibit the removal of surface material, the more so in the absence of lubrication. This is called wear, and may enhance fatigue and corrosion. A wear resistant surface is generally hard, tough, and smooth.

9. **Elastic Modulus.** The ratio of stress to strain demonstrated by the material when loaded so as to cause no permanent deformation.

10. **Density.** Material mass per unit volume.

11. **Coefficient of Thermal Expansion.** The fractional change in length occurring per degree of temperature change.

12. **Thermal Conductivity.** The speed with which heat energy can travel through the solid material.

13. **Oxidation and Corrosion Resistance.** Oxidation is the formation of oxides on the surface of the specimen; with superalloys it usually results in the formation of a tough oxide scale. This scale periodically flakes off the surface in a process called spalling, resulting in a net loss of material from the specimen. When present, the oxide scale can significantly increase the resistance to further oxidation.

Aqueous corrosion is corrosion due to the presence of water in liquid or vapor form; it is active at ambient temperatures but not at engine operating temperatures.

Hot-corrosion covers a number of corrosive mechanisms and involves oxygen as well as other elements in the hot gas—for gas tubines these elements are typically vanadium, sulphur, sodium, potassium, and chlorine arising from the fuel and the ambient atmosphere.

Oxidation and corrosion may occur on the material surface and in metal grain boundaries; when they occur in the latter they may have a deleterious effect on tensile strength, creep and fatigue life, and toughness. A protective oxide scale can be removed by corrosive action, thereby increasing the oxidation rate. Hot-corrosion of superalloys is most rapid at intermediate temperatures--in the hot-corrosion window of 1200-1700F--while oxidation rate generally increases with temperature.

14. **Burnout Temperature.** As the temperature of the hot gas flowing past a specimen increases beyond a certain level, the specimen will begin to lose material from its surface by rapid oxidation. The temperature at which the material loss rate exceeds a given high value is called the burnout temperature, and it approaches the melting temperature of the material. Burnout temperature is generally dependent upon the material, the specimen geometry, and the pressure and speed of the gas flow.

When dealing with non-metallic hot-section materials (such as ceramics) there are several important properties in addition to those described above. These are:

15. **Weibull Modulus.** Non-metals, particularly ceramics, often exhibit a wider scatter in properties than do metals. Consequently, properties such as strength are often described in statistical terms. The Weibull modulus for a given type of material describes the strength distribution of that material. The Weibull modulus may be given for a number of different strengths (tensile, shear, stress-rupture, fracture toughness, fatigue) although it usually refers to ultimate tensile strength.

16. **Static Fatigue.** This is a rather confusing term which means degradation in static strength with time. Such degradation may occur due to oxidation or due to other microstructural changes in the material.

Major Property Requirements of Specific Parts

While many material properties are desirable to some degree in most hot-section parts, a small subset of them may be identified as having primary importance for any given part. These major property requirements have been identified by the industry for all gas turbine engine parts--from the service environment of the part and from service experience (particularly failure modes). Within the constraints of cost and materials availability, these major property requirements drive the design selection of materials to be used in the engine. The major property requirements of several hot-section components are listed in Table 2 and briefly discussed below; these requirements apply to metal components.

1. **Combustor liner.** This sheet-material part is heated by both the radiation from the naked flame and by the hot turbulent gases; it is cooled by incoming air from the compressor. The first materials property requirements are a high burnout temperature and high thermal fatigue strength, followed by good resistance to oxidation and hot-corrosion, and creep resistance. Buckling due to creep leads to high spots that are readily burned through. The liner material must be able to be easily fabricated (by rolling into sheet and usually welding) and repaired (welding and re-coating).

2. **Turbine vane.** The various vanes are exposed to very high temperatures and to the full velocity of the hot gases passing through the turbine. These conditions are most severe for the first stage vanes, generally

Table 2
Aircraft Gas Turbine Engine Metal Component
Material Property Requirements and Failure Modes

Component	Major Material Property Requirements*	Typical Failure Modes
Combustor liner	High burnout temperature Thermal fatigue strength Oxidation & hot-corrosion resistance Creep resistance Fabricability Repairability	Thermal fatigue Creep-buckling Burn-through HCF
Turbine vane	High burnout temperature Fracture toughness Thermal fatigue strength Oxidation & hot-corrosion resistance Erosion resistance Low distortion Repairability	Thermal fatigue Oxidation Hot-corrosion Distortion (creep)
Turbine blade	High burnout temperature Tensile strength (yield & ultimate) Stress-rupture life Creep resistance Oxidation & hot-corrosion resistance Fatigue strength (HCF & LCF) Fracture toughness Erosion resistance Low thermal expansion High thermal conductivity Low density Repairability	Stress-rupture Oxidation Hot-corrosion Thermal fatigue HCF
Turbine disc	Rim: Stress-rupture life Creep resistance Fatigue strength (LCF) Oxidation & hot-corrosion resistance Hub: Tensile strength (yield & ultimate) Fatigue strength (LCF) Low density for entire disc	Mechanical LCF HCF
Cases	Tensile strength (yield & ultimate) Stress-rupture life Creep resistance Fatigue strength (LCF) Oxidation & hot-corrosion resistance Low distortion in blade region Fabricability Repairability	LCF (mostly mechanical) HCF

*Listed in approximate order of decreasing priority for each component.

Source: OTA

diminishing progressively downstream (Fig. 6). A key design consideration for first stage vanes is gas temperature variations along the length of the vane. First stage vanes are usually air-cooled and the later stage vanes may be also. Since vanes are stationary they are not highly loaded. However, thermal stresses may be large; and since the vane is an airfoil that turns the flow, the pressure forces over its surface produce moderate bending moments and stresses in the vane. A high burnout temperature is required especially for first stage vanes. Fracture toughness is very important to maintain integrity in the event that foreign objects (such as stones, birds, and perhaps failed compressor parts) pass through the turbine. Thermal fatigue strength, and oxidation and hot-corrosion resistance are vital. Erosion resistance is important, especially for first stage vanes where gas velocities are highest. Low distortion in service is necessary to preserve design flow conditions. Finally, repairability is vital for maintaining low engine life-cycle costs--the vanes should be weldable to repair cracks and able to be re-coated, unless an alloy is selected that will last for the design life of the engine.

3. **Turbine blade.** The turbine blade presents the most difficult materials problem in the gas turbine engine; the problems are less severe for the low pressure turbine than for the high. While the temperature is slightly less than that experienced by vanes in the same stage, the centrifugal loads on the blades due to turbine rotation are very large: a blade weighing only several ounces may experience a load of two or three tons on its firtree attachment. A high burnout temperature is required. High tensile srength (both yield and ultimate) is necessary because of the centrifugal loads. The combination of high temperature and large radial stress requires high stress-rupture strength; further, it is desirable to maintain low clearances at the blade tips in order to minimize gas losses and this requires high creep resistance. Oxidation and hot-corrosion resistance are very important. High-cycle fatigue strength is important for the entire blade due to the high frequency gas loads imposed (the blades pass close to the vanes). Low-cycle fatigue strength is important also--primarily mechanical LCF occurs in the firtree region and thermal LCF in the airfoil region. A low modulus of elasticity in the radial direction, good ductility, and high thermal conductivity contribute

significantly to good thermal fatigue resistance. Fracture toughness is more important for blades than for vanes because blade failure causes turbine imbalance that may lead quickly to catastrophic failure of the whole turbine. Erosion resistance is important but not as much as for vanes. A low coefficient of thermal expansion is required to minimize blade tip clearance variations. Low density reduces the centrifugal loads imposed at a given turbine speed. Finally, repairability is important--particularly the ability to be re-coated (weld repair is usually not possible). Achieving all of these properties together results in very high costs for turbine blades.

4. **Turbine disc.** The materials properties required at the rim of the disc where the blades are attached are different from those required in the hub or central region of the disc. The rim operates at high temperature (around 1300F in some engines), has a complex firtree geometry for blade attachment (Fig. 10) and supports the radial and side loads of the blades. There are both radial and hoop stresses in the disc--a strong multidirectional stress field as opposed to the nearly unidirectional stress field in the blades. The required material properties at the rim are stress-rupture life, creep resistance to prevent distortion of the firtree, low-cycle fatigue strength (both mechanical and thermal), and moderate oxidation and hot-corrosion resistance. In the hub region, the operating temperature is significantly reduced (perhaps 400F below rim temperature), the stress field is more intense, and there are no oxidation and hot-corrosion problems. Therefore in the hub region the required material properties are tensile strength (both yield and ultimate--the latter is often referred to as burst strength), and mechanical and thermal low-cycle fatigue strength. A low density disc is always desirable, for this reduces the stresses imposed by rotation.

5. **Cases.** Hot-section cases generally require all the properties required of discs, with a few additions (low density is important to reduce engine weight but not stresses). The additions are: low distortion in the vicinity of the blade tips in order to preserve design clearances; fabricability, particularly weldability; and welding repairability.

These various parts typify the materials requirements for gas turbine hot-section parts. The turbine blade and disc present the most demanding requirements.

HOT-SECTION MATERIALS

In this section we discuss the materials that have been evolved to satisfy the property requirements of hot-section parts. It is intended to show the unique position the superalloys have held in fulfilling those requirements, and the composition and structure of some typical superalloys and coatings. It concludes with a discussion of several new materials that may eventually replace the superalloys in their role as hot-section materials.

Background

To place the hot-section materials in perspective, we first briefly review the materials that have been and now are used in the cooler parts of the engine (1,5-8).

1. **Compressor.** Rotors in low-pressure (LP) and high-pressure (HP) compressors were initially made from low alloy steels, and sometimes plated with cadmium (LP) or cadmium plus nickel (HP). Titanium was introduced in the mid 1950s and was subsequently employed for LP rotors. As engine performance has increased, and as the temperature capability of Ti alloys has increased, these alloys have been used for HP compressor rotors in some engines; other engines have resorted to nickel-based superalloy hp compressor rotors. Blades and vanes for both LP and HP compressors were initially made from 12% Cr stainless steel; this usage has continued for vanes. More recently, Ti alloys have been used for LP blades. High-pressure compressor blades have also used Ti alloys; but for the higher temperatures of modern engines, stainless steel or nickel-based superalloys are used.

2. **Fan.** "Fan engines would not have been possible without the availability of titanium alloys" (5). Fan rotors, blades, and vanes are all fabricated from these alloys. Recently fan blades have employed graphite or boron fiber-reinforced aluminum or polymer; and evolving high strength aluminum alloys and graphite fiber-reinforced polymers have been used for fan vanes.

3. **Shafts.** These have always been made of steel, sometimes variously coated or pl ed for corrosion resistance. Some shafts are made from superalloys.

4. **Cases.** Fan cases have used stainless steel, and titanium and aluminum alloys. Compressor cases have used stainless steel, and this usage continues; titanium is also used.

Now consider the hot-section. The materials generally used for the cooler engine parts--stainless steel, titanium, aluminum, and various composites--are not suitable for the hot-section, either because their melting temperature is too low or because they retain insufficient strength at operating temperatures. To fulfill the property requirements of the hot-section, the **superalloys** were developed. The development of turbine blade and vane alloys is briefly reviewed below (9).

Gas turbine blade and vane materials have evolved from the materials used in the British Whittle engine, and from those used in the U.S. for turbosuperchargers* in the 1930s. The first Whittle engine used stainless steel turbine blades, but these were quickly replaced by a 75% nickel-25% chromium alloy. An improved version of this alloy, strengthened by Al and Ti, was designated Nimonic 80A (Table 3 and Fig. 14); the British subsequently developed a series of increasingly complicated and stronger nickel-based Nimonic alloys. The addition of Al and Ti to the basic Ni-Cr alloy was the fundamental step taken in the development of the Ni-based superalloys. Such additions result in the precipitation of small particles of a metal phase called gamma prime (γ') which is a combination of Ni, Al, and Ti: $Ni_3(Al, Ti)$. It is the γ' phase that gives the Ni-based superalloys their high temperature strength.

The first blades and vanes for U.S. turbosuperchargers were wrought iron-based alloys of which alloy 17W was typical. Proving too weak, these alloys were replaced in World War II--first by the Ni-based alloy Hastelloy B. The first U.S. jet engine, the General Electric I-A, flew with Hastelloy B blades in 1942. However, due to forging difficulties with this alloy and the shortage of forging equipment, a change was made in turbosupercharger blades to the cast cobalt-based alloy HS21.

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* A turbosupercharger uses a turbine to extract shaft power from the hot exhaust gases of a piston engine; it uses that power to compress the air being inducted into the engine.

Table 3
Historical Development of High-Temperature Alloy Compositions

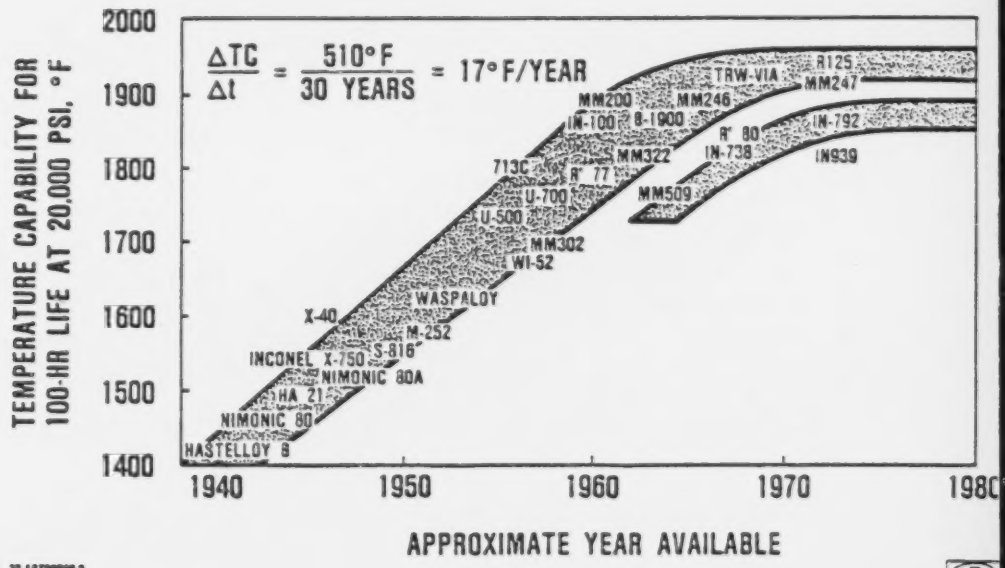
Alloy	Composition (% by weight)										
	C	Cr	Mo	Co	W	V	Fe	Ni	Al	Ti	Cb
Nimonic 80A	—	20	—	—	—	—	—	76.6	1.2	2.2	—
17W	0.50	12	1.0	—	2.2	—	65.3	19	—	—	—
Hastelloy B	—	—	28.0	—	—	0.3	5.0	66.7	—	—	—
HS21	.25	27	6.0	64.75	—	—	—	2	—	—	—
Tinidur	—	14.7	—	—	—	—	56.8	26	0.2	2.3	—
A-286	—	15	1.3	—	—	0.3	55.0	26	0.3	2.1	—
X-40	.40	25	—	56.6	8.0	—	—	10	—	—	—
M-252	.15	19	10.0	10	—	—	—	57.35	1.0	2.5	—
MAR-M200	.15	9	—	10	12.5	—	—	60.35	2	5	1

Key:

C	Carbon	Fe	Iron
Cr	Chromium	Ni	Nickel
Mo	Molybdenum	Al	Aluminum
Co	Cobalt	Ti	Titanium
W	Tungsten	Cb	Columbium (Niobi
V	Vanadium		

Source: Ref 9.

FIG. 14 TEMPERATURE CAPABILITY OF SUPERALLOYS



Source: Garrett Turbine Engine Company

The first German jet fighter employed Tinidur, an iron-based alloy, for its turbine blades because of a shortage of Ni and Co in that country during World War II. Tinidur was subsequently modified in the U.S. to become the very successful alloy A-286, an FeNi-based superalloy.

Alloy development continued in the U.S. during World War II. During the initial period following the war the strongest turbine blade and vane alloys were Co-based, a representative alloy being cast X-40. Then a significant change occurred.

> The advent of vacuum induction and consumable electrode vacuum arc melting in the early 1950's reversed the trend of turbine airfoil materials from cobalt to nickel-base alloys. The prime precipitation hardening elements in these alloys had been aluminum and titanium since the days of "Nimonic 80A" and "Inconel W," both of which were developed in the 1944-1945 period. The conventional air-melting practices used on these alloys made it difficult to retain large amounts of titanium and aluminum to produce greater strength, and ingots were often inhomogeneous and plagued with inclusions. Vacuum melting circumvented these problems, and a whole new family of nickel-base high temperature alloys were spawned (9).

A typical alloy in this new nickel-based family was M-252 (developed in 1951), being both stronger and lighter than the then strongest Co-based alloy.

Since then further strength improvements have been made in turbine airfoil Ni-based alloys, allowing growth in use temperature, turbine speed, and hence airplane performance (Fig. 14). These strength improvements have been made in three ways:

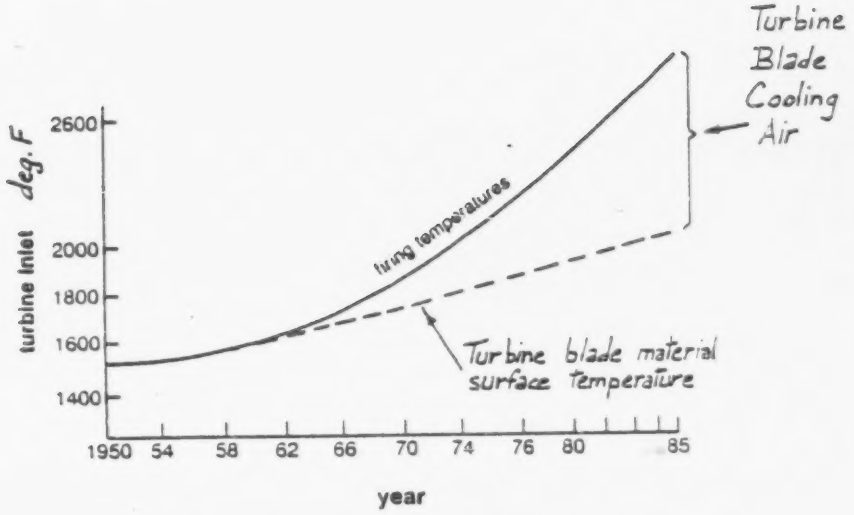
1. The quality of conventional castings has been improved, allowing turbine airfoil manufacturers to take advantage of the fact that both Co- and Ni-based alloys have greater stress-rupture strength at high temperature in cast rather than wrought form. Until the early 1960s, wrought blades were preferred because of casting quality problems.
2. The Ni-based alloys have been strengthened by minor modifications to their composition and heat-treatment. Composition changes have involved increases in the composition of Al and Ti; variations in Co content; variations in the composition of the refractory metals Cb, Mo, Ta, and W; and small variations in B, C, V, Zr, Hf, and Re.
3. Advances in casting technology have led to the development of

Directional Solidification. This has permitted the evolution of fabrication technologies using directionally solidified (DS) polycrystalline alloys and single crystals (SC). Their directional lattice structure and the absence of transverse grain boundaries give directionally cast turbine blades and vanes (DS and SC) higher stress-rupture and thermal fatigue strengths than those conventionally cast. The first successful DS turbine blade was introduced by Pratt & Whitney Aircraft in 1969 and was made from the alloy MAR-M200 (an alloy developed by the Martin Marietta Corporation); the composition of this alloy (Table 3) is representative of DS and SC Ni-based turbine alloys in use today.

The strength and temperature capabilities of Co-based turbine airfoil superalloys have not markedly increased since the advent of alloy X-40. The Co-based alloys do not contain the γ' phase, and cannot be heat-treated. The small performance improvements that have been achieved with these alloys are attributable to minor variations in alloy composition and conventional casting quality.

In summary, successful developments of turbine airfoil materials to date have evolved a series of Co-based, Ni-based, and FeNi-based superalloys. The Ni-based alloys are the strongest and are used for turbine blades; the Co-based alloys are used along with the Ni-based alloys for turbine vanes. The FeNi-based alloys such as A-286 are no longer used for airfoils, but are used for combustors, discs, and cases. Further developments of turbine blade and vane materials are in progress, namely Rapid Solidification Processed Superalloys, Long Range Order alloys, Eutectics, Oxide Dispersion Strengthened Superalloys, Fiber-Reinforced Superalloys, ceramics, and carbon-carbon composites. These technologies are in various stages of development and will be discussed in a later section.

Materials for hot-section parts other than turbine blades and vanes have evolved in a fashion similar to that described above. Combustor liners are made from Co- or FeNi-based superalloys in sheet form; to facilitate rolling, the FeNi-based alloys used contain none of the precipitated γ' phase. Turbine discs are forged from Ni-based or FeNi-based alloys--discs have much higher tensile strength when forged than when cast. For some lower temperature



Trends showing improvement in firing temperature.

FIG. 15 TURBINE INLET TEMPERATURE TRENDS

SOURCE: REF. 4 and OTA

applications, turbine discs have been forged from stainless steel or titanium. Discs have begun to make use of new Powder Metallurgy technologies, since some of the very high strength alloys cannot be successfully forged from conventional ingots but can be formed by the hot-isostatic pressing or hot-forging of metal powder. Small ceramic turbine discs and integrally formed disc/blades are being developed for automotive and unmanned aircraft (missile) gas turbine use, but will not see manned aircraft service for many years. Cases for combustors, turbines, jet pipes, and nozzles were initially made from stainless steel. As temperatures have increased and alloys improved, these parts have made increasing use of FeNi- and Ni-based superalloys.

In parallel with all these materials developments, two other important technologies have evolved: cooling technology for turbine airfoils, and coating technology for turbine airfoils and combustor liners.

1. Developments in turbine airfoil cooling technology have significantly increased the permissable turbine inlet temperatures for the airfoil materials in use. Moreover, the quantity of cooling air required to perform this task has steadily been reduced due to improving cooling efficiency; this reduction in turn minimizes the engine performance losses inherent in turbine cooling. Figure 15 shows the increases in TIT that have occurred since 1950 and made possible by materials improvements and cooling improvements. The TIT increase due to materials has been about 17F per year; this has yielded an increase in engine thrust of 1.5% to 2.0% per year and an increase in engine efficiency of 0.3% to 0.6% per year (4).
2. Three basic types of coatings have evolved: metallic diffusion coatings; metallic overlay coatings; and thermal barrier coatings. Diffusion and overlay coatings have been used on turbine blades and vanes and on combustor liners. Thermal barrier coatings have been used on combustor liners and on turbine vane platforms.

Typical Current Structural Alloys and Coatings

The preceeding sections have shown that hot-section materials must demonstrate a wide range of very demanding properties and that the search for such materials has resulted in the evolution of three families of structural metal alloys: the Ni-based, FeNi-based, and Co-based superalloys. Each of

these alloys may be distinguished from the others in two ways: composition and form.

1. The composition is given by the weight percent of each alloying element present, including: Ni, Cr, Co, Fe, Al, Ti, Cb, Mo, Ta, W, B, C, V, Zr, Hf, and Re.
2. The form is defined by the alloy macrostructure and microstructure. The macrostructure is described by grain size and orientation; the microstructure by the metal phases present and by the distribution of those phases in the grains and grain boundaries. Both the macrostructure and microstructure depend upon the process history of the alloy--conventional casting, directional solidification, forging, rapid solidification of powder and subsequent consolidation by hot-isostatic pressing or forging, heat treatment, and so on.

The designer chooses the composition and form of an alloy in order to achieve the special mix of properties required for a given component. Some of the numerous structural alloys currently used for hot-section parts are given in Table 4 where a number of trends may be observed. Those trends are discussed below. The alloys shown may be variously coated; coatings currently in use are described at the end of this section.

The two most used combustor liner alloys are shown in the table. They are FeNi- and Co-based respectively, with a 22% Cr content; Co content is 1.5% and 41% respectively. Both alloys are used in large grained sheet form.

The turbine vane alloys are shown in decreasing order of nominal operating surface temperature. The Oxide Dispersion Strengthened (ODS) superalloy MA-754 has the highest temperature capability and contains no cobalt; ODS superalloys are not yet widely used and are discussed in the next section. The DS Ni-based alloy MAR-M200 is next with 10% Co. Of the conventionally cast alloys, the Co-based alloys generally have the highest temperature capability--note that these alloys contain more Cr than any of the other vane alloys. All of the cast vane alloys are used in large grained form. Single crystal first stage turbine vanes are being qualified by several engine manufacturers and should be in production in the near future.

Table 4
Typical Structural Alloys Used for Hot-Section Components

Component	Alloy	Composition (% weight) Ni	Co	Cr	Fe	Form	Nominal Operating Conditions: Stress	Surface Temp. F
						--Large Grains--		
Combustor Liner	Hastelloy X	48	1.5	22	18.5	Sheet	Low	1600
	HA-188	22	41	22	—	Sheet	Low	1600
						--Large Grains--		
Turbine Vane	MA-754	78	—	20	1	ODS+HIP, ODS+forge		1900+
	MAR-M200	60	10	9	—	DS		1900
	MAR-M247	60	10	8	—	CC, DS		1800-1850
	MAR-M509	10	55	23.5	—	CC	Moderate	1800+
	X-40	10.5	56	25.5	—	CC	to	1800+
	IN-713	72.5	—	13.5	—	CC	Low	1600
	Rene-77	55	15	15	—	CC		1600
						--Large Grains--		
Turbine Blade	Alloy 454	62.5	5	10	—	SC		1900
	MAR-M200	60	10	9	—	DS		1850
	MAR-M247	60	10	8	—	CC, DS		1800-1850
	B-1900	65	10	8	—	DS	High	1800
	Rene-80	60.5	9.5	14	—	CC		1750
	IN-713LC	72.3	—	12	—	CC		1600
	Rene-77	55	15	15	—	CC		1600
						--Small Grains--		--Rim Temp.-
Turbine Disc	IN-100	56	18.5	12.5	—	PM+HIP, PM+forged		1300
	MERL-76	54.1	18.5	12.4	—	PM+HIP, PM+forged		1300
	Astroloy	55.5	17	15	—	PM+HIP, PM+forged, forged		1300
	Waspaloy	58	13.5	19.5	—	Forged		1250
	Rene-95	61.3	9	14	—	PM+forged	High	1200
	IN-718	53	—	19	18	Forged		1200
	IN-901	45	—	12.5	34	Forged		1100
	A-286	25.5	—	15	55	Forged		1000
						--Small Grains--		
Case	Waspaloy	58	13.5	19.5	—	Sheet, forged		1300
	IN-718	53	—	19	18	Sheet, CC, forged	High	1200
	IN-901	45	—	12.5	34	Forged		1100
	A-286	25.5	—	15	55	Forged		1000

CC = Conventionally cast
DS = Cast and directionally solidified polycrystal
SC = Cast and directionally solidified single crystal
PM = Powder metallurgy
ODS = Oxide dispersion strengthened
HIP = Hot-Isostatic Pressing

Alloy base compositions are shown in bold face type.

Source: OTA

The most common turbine blade alloys are all cast, large grained Ni-based superalloys. Their temperature capability decreases from single crystal through directionally solidified to conventionally cast. Note that as the temperature capability drops (indicating a decrease in alloy strength) the Cr content increases. The lower Cr alloys are more dependent upon coating integrity than higher Cr alloys.

Turbine discs are wrought from Ni-based or FeNi-based alloys, are fine grained, and operate at temperatures considerably lower than the blade alloys. Hub temperatures may be 400F cooler than the rim temperatures shown. The strongest disc alloys employ powder metallurgy technology, being either forged or hot-isostatically pressed. Of the alloys forged from conventional ingots, Waspaloy is the most used--it is less expensive than the PM alloys and being Ni-based is stronger than the FeNi-based alloys. Waspaloy is the biggest consumer of Cobalt in discs. The Cr content of the disc alloys is generally higher than that of the blade alloys, usually being in the range 12-20% Cr.

The case alloys shown in the table are the same as the last four disc alloys. Waspaloy has the highest temperature capability and is the only one shown containing Co; it also contains the most Cr.

There are some important connections between macrostructure and superalloy material properties which explain the use of the various macrostructures shown in Table 4 (cf. the material requirements in Table 2).

1. Large grains, regardless of their lattice orientation, give increased stress-rupture life and reduced creep rate at high temperatures. Stress-rupture failure is intergranular (except for DS and SC); the stress-rupture life is extended and the creep rate reduced by reducing the number of grain boundaries. Large grains are therefore used for combustor liners, and turbine vanes and blades.
2. Small grains give high tensile strength (yield and ultimate) and high fatigue strength at moderately high temperatures. At high temperatures grain size has little effect on these properties. Small grains are used for discs to achieve the required hub properties, thereby somewhat reducing the creep resistance of the rim region which

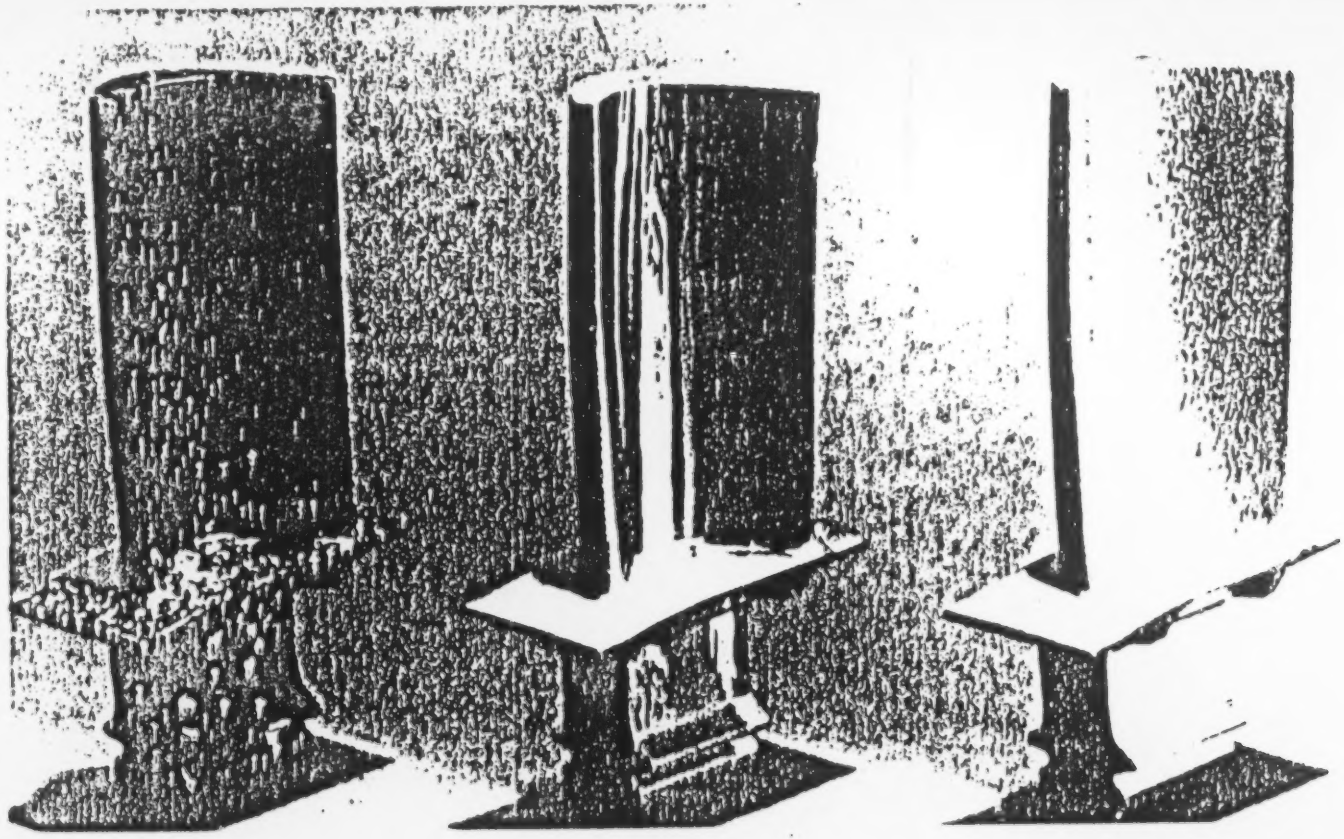
would best utilize large grains. Cases employ small grains to optimize tensile strength and fatigue strength.

3. Powder metallurgy (PM) has enabled the introduction of the highest strength disc alloys (10). Rapid solidification of the powder permits greater alloying freedom, and the small grains that result from the use of powder easily slide over one another during forging. In the absence of the fine grains that PM makes possible, these strong alloys could not be forged into discs.

4. Directional solidification results in long parallel grains whose lattices run parallel to their length. A number of these grains are used to make a directionally solidified (DS) turbine blade or vane; the long columnar grains are grown parallel to the blade or vane length (Fig. 16). DS blades have a lower longitudinal modulus of elasticity than occurs with the randon (equiaxed) orientation of grain lattices in conventionally cast blades. This reduced modulus enhances thermal fatigue life. Further, the longitudinal grain boundaries are not highly stressed in a DS blade, and so stress-rupture life is increased and creep reduced compared to a conventionally cast blade. All this permits higher temperature operation--DS turbine blades can operate at temperatures about 50F above conventionally cast blades.

5. Single crystals are employed to make turbine blades (Fig. 16). They are formed by a directional solidification process wherein only one crystal (grain) survives to form the blade--this lone crystal has its lattice running parallel with the blade length. Single crystal blades give better performance than DS blades (11). This is mainly because of the absence of any grain boundaries. It is also because certain grain boundary control elements (B, C, Hf, Zr) that are used in polycrystalline structures--and which are melting point depressants--can be removed from single crystal alloys thereby enabling heat-treatment at higher temperatures. A single crystal blade with modified grain boundary control elements has greater stress-rupture life, and greater high-cycle and low-cycle fatigue life than a DS blade of the same (unmodified) alloy. The elimination of grain boundaries also improves corrosion resistance. The superior properties of single crystal blades permit them to operate at temperatures about 50-75F higher than DS blades.

ADVANCES IN TURBINE AIRFOIL MATERIALS



Conventional casting | Columnar grain | Single crystal

FIG. 16 Turbine blade macrostructures

Source: United Technologies Pratt & Whitney Aircraft

Coatings. The surface properties (oxidation, hot-corrosion, erosion, and wear resistance) of the structural alloys are often unsatisfactory; various coatings are used to enhance these properties. Three types of coatings are currently in use: diffusion coatings; overlay coatings; and thermal barrier coatings (12). The latter provide some thermal insulation for the underlying structural material (substrate), as well as enhancing oxidation and hot-corrosion resistance. These various coatings make an important contribution to the design life of many components.

Diffusion coatings were first employed in the early 1960s to enhance the oxidation and hot-corrosion resistance of turbine blades. These consisted of surface enrichment with Al and/or Cr; diffusion of these elements into the substrate formed the protective layer. During service, slow diffusion of the same elements back to the surface provided Al for the formation and renewal of a protective aluminum oxide scale (for oxidation resistance) and Cr for hot-corrosion resistance. These coatings are still widely used today for turbine blades and vanes--Al diffusion coatings are the most common (Table 5).

Overlay coatings are more recent and less widely used. As the name suggests, these overlay the substrate rather than diffusing into it. Some of these coatings are MCrAlY alloys where M stands for Ni, Co, and/or Fe. These coatings are used on turbine blades and vanes (Table 5). Diffusion and overlay coatings can extend component life by a factor of ten.

Metallic coatings have also been used for enhancing wear resistance. These too are overlay coatings, being typically "surface deposits of cermets consisting of chromium, tungsten, or tantalum carbides in a cobalt matrix" (12) applied by thermal spray, fusion welded hardface alloys, or electroplated Cr.

Thermal barrier coatings (TBCs) were first introduced in the late 1960s to extend the life of combustor liners. TBCs comprise a metallic overlay coating (such as an MCrAlY coating) which is in turn coated by a thin layer of ceramic oxide. The metallic coating is referred to as a bond coat. The ceramic coatings used are typically zirconia-based. TBCs are currently used on combustor liners and turbine vane platforms (Table 5). The TBC provides

Table 5
Typical Coating Materials for Hot-Section Components

Coating Type	Coating Composition (% weight)	Component Coated	Coating Technique
Diffusion	• Al (mostly used) • Al then Cr	Major coating used for turbine blades and vanes	1. Pack cementation process plus heat-treatment to diffuse coating 2. Slurry 3. Vapor deposition
Overlay (MCrAlY)	• Ni-21Co-19Cr-12Al-0.3Y (mostly used)	Turbine blades and vanes	1. Physical vapor deposition by electron beam melting (EB/PVD) 2. Plasma spray
	• Ni-18Cr-12Al-0.3Y • Co-20Cr-12Al-0.3Y	Turbine blades and vanes	
Thermal Barrier	Bond coat: • Ni-20Cr • Ni-12Al • Ni-21Co-19Cr-12Al-0.3Y	Combustor liners	Plasma spray
	• Ni-18Cr-12Al-0.3Y • Ni-21Co-19Cr-12Al-0.3Y	Turbine vane platforms	
	Ceramic: • ZrO_2-21MgO	Combustor liners	Plasma spray
	• ZrO_2-6Y_2O_3	Turbine vane platforms	

Source: OTA

oxidation and hot-corrosion resistance by means of the bond coat, and thermal insulation by means of the ceramic layer. A ten mil thick ceramic layer can lower the substrate temperature about 10C or 18F; reductions in substrate temperature of 30F have been achieved (12). These temperature reductions can result in substantial increases in the stress-rupture and fatigue lives of the substrate alloy. Extension of TBC use to turbine airfoil surfaces is a possibility for the late 1980s (13). These applications can reduce cooling air requirements and substantially increase turbine blade and vane life; they also provide the opportunity for significant increases in turbine inlet temperature and hence engine performance. TBCs might be employed on combustor liners and turbine airfoils in order to allow the use of substrate alloys with lower strategic element content and having lower temperature capability. TBCs have not yet been successfully applied to turbine airfoils because their tendency to spalling has not yet been overcome (spalling is local peeling of the coating from the substrate). TBCs do however offer a cost-effective way of improving performance or blade/vane life--on a conventionally cast blade such a coating raises the cost of the finished blade by only 20% approximately (14).

Materials Under Development

The hot-section materials described above are currently in service in military and commercial gas turbine engines. To meet the demand for increases in operating temperature, several additional structural materials and their manufacturing processes are receiving a great deal of research and development effort. These materials, which are at various stages of development, are described below; they are targeted mainly at turbine blade and vane usage, but some of them have other potential uses. Estimates of the nominal operating surface temperatures of turbine blades fabricated from some of these materials and of their approximate manned aircraft service introduction dates in this application have been made by the National Aeronautics and Space Administration (NASA)--these estimates are shown in Figure 17.

1. **Rapid Solidification Processed Superalloys.** These materials are formed by the consolidation of rapidly solidified metal powder, flake, or ribbon. A number of rapid solidification processes have been developed--e.g., inert gas atomization, vacuum melt spinning, and the Pratt & Whitney Aircraft

TURBINE BLADE MATERIALS TRENDS

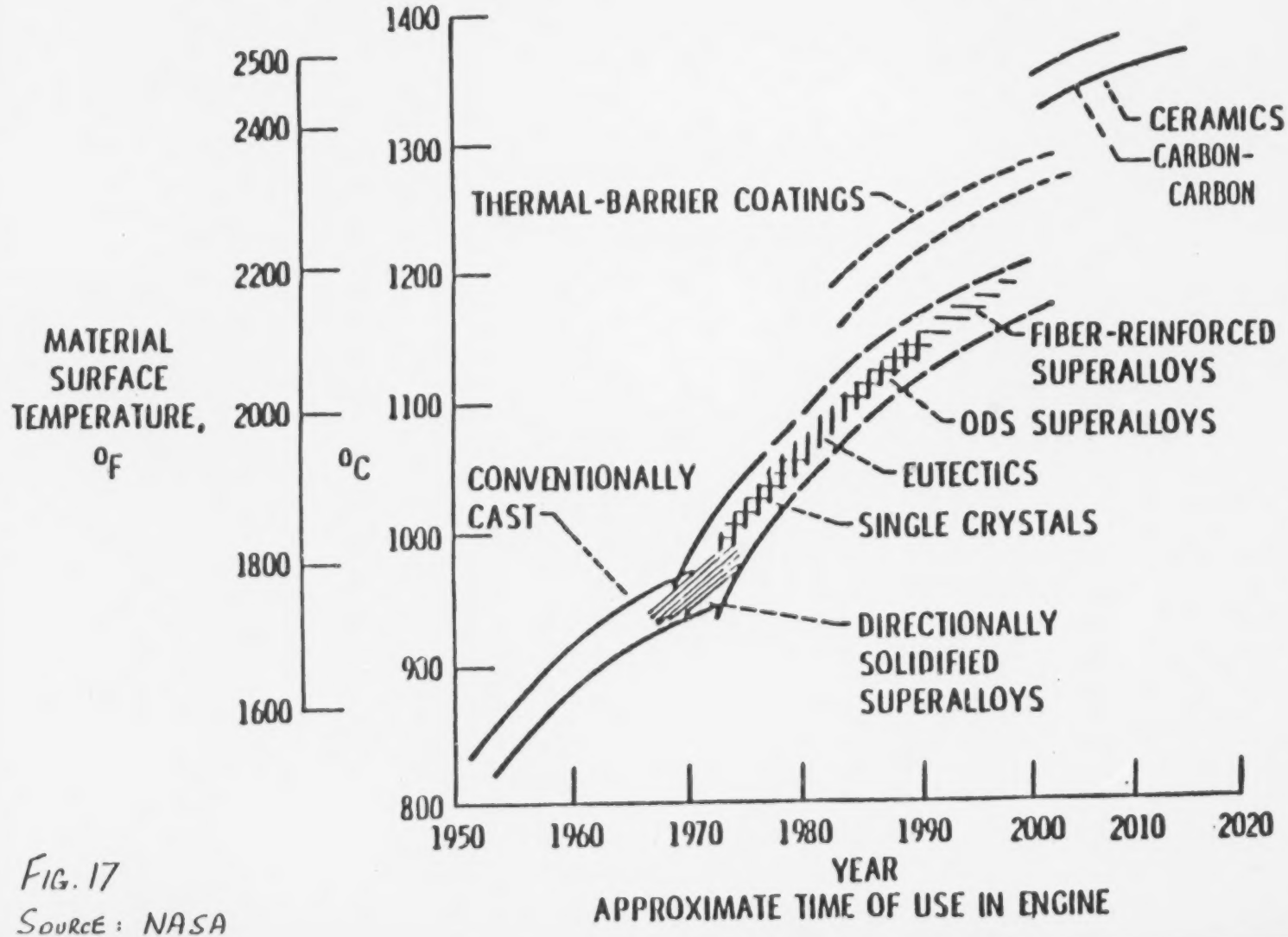


FIG. 17

SOURCE: NASA

CREEP DEFORMATION OF SUPERALLOYS AT 1800°F AND 30,000 psi

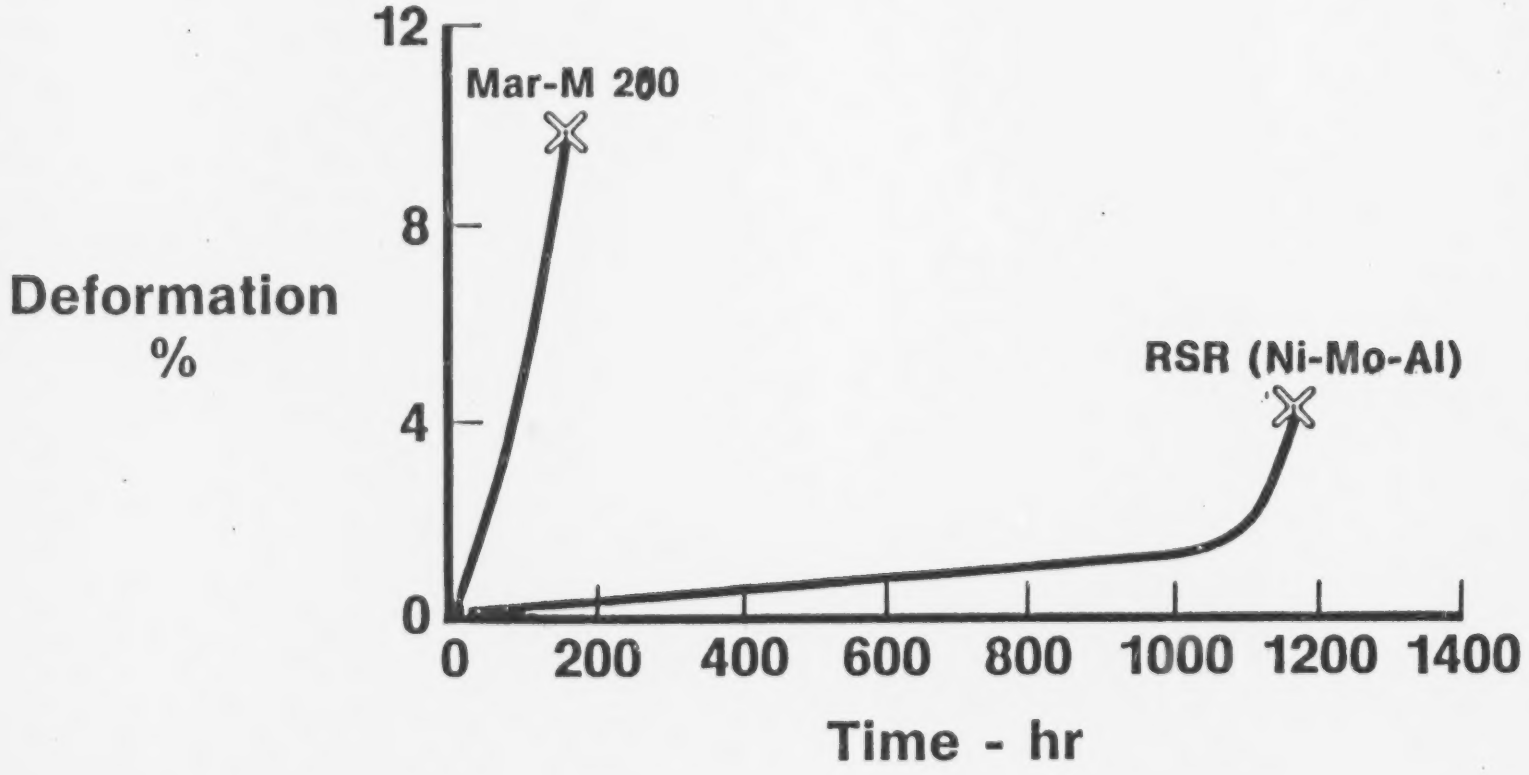


FIG. 18
SOURCE : REF. 15

OXIDATION RESISTANCE OF SUPERALLOYS

CYCLIC TESTING AT 2100°F

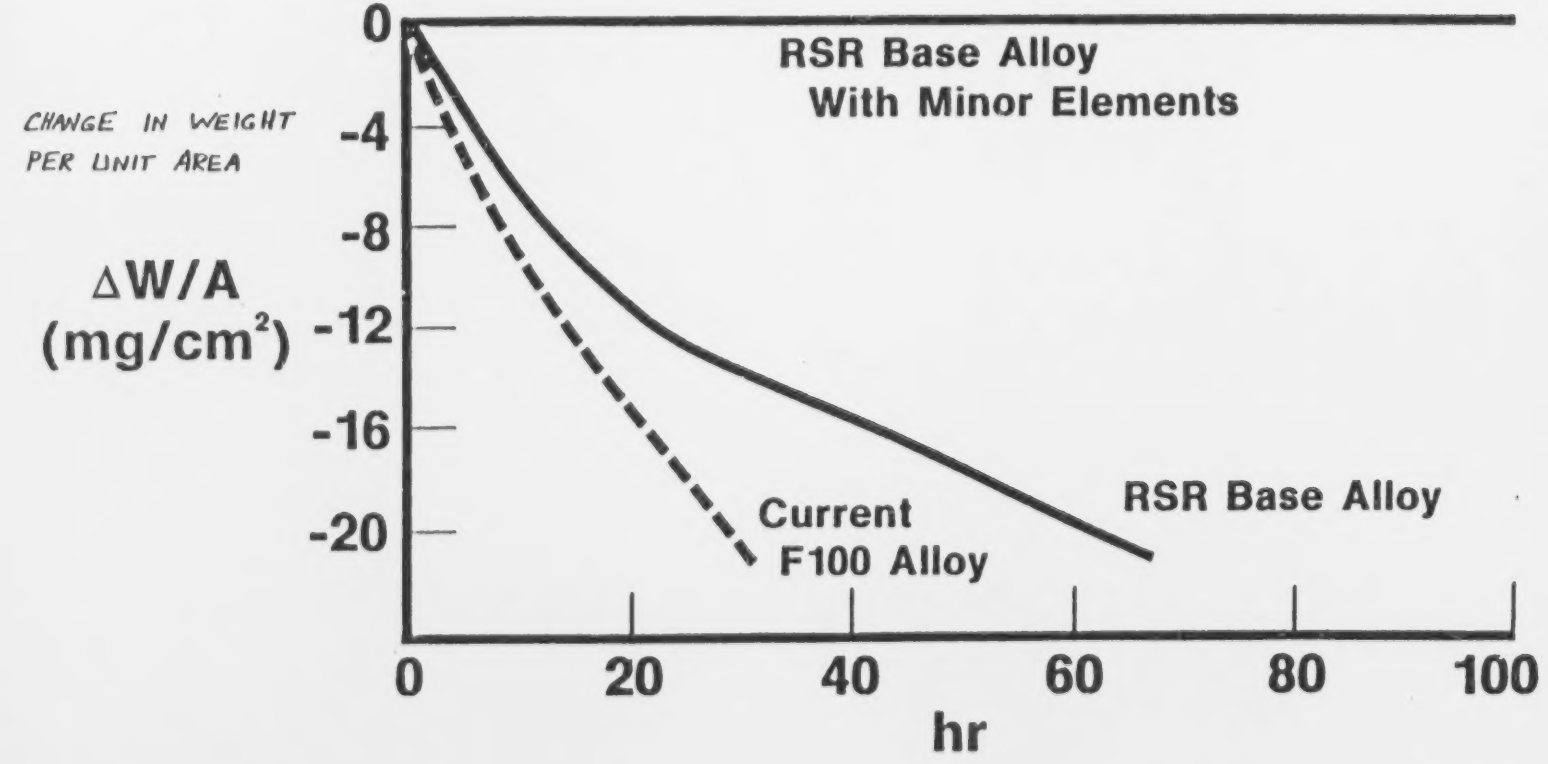


FIG. 19
SOURCE: REF. 15

Rapid Solidification Rate (RSR) process (11,15). With such processes, solidification rates in excess of 100,000 deg.C/second are possible (compared with conventional solidification rates of less than 1 deg.C/sec). The resultant powder, flake, or ribbon is consolidated by various methods including extrusion, hot-isostatic pressing, and forging. These processes result in fine, equiaxed grains suitable for discs and shafts. Machining is performed as desired. For use in combustor liners or in tubine blades and vanes, the fine grained product must be heat-treated in a process called directional recrystallization in order to develop the required large, directionally structured grains.

Rapid solidification leads to a far more homogeneous distribution of alloying elements (in the powder, flake, ribbon and consolidated/heat-treated product) than do conventional solidification rates. As a result, components made using rapid solidification of an alloy designed for conventional processing have improved properties compared with those made of the same alloy conventionally processed. Further, rapid solidification extends alloying flexibility thereby enabling the development of otherwise unattainable alloys with even better properties (11,16).

As an indication of the potential that rapid solidification offers for the development of new hot-section alloys with superior properties, consider the RSR NiMoAl ternary that Pratt & Whitney Aircraft is developing for turbine airfoils (15). Figure 18 compares the high temperature creep of this alloy with that of cast DS MAR-M200; the RSR ternary here contains no Co or Cr. Figure 19 shows the oxidation resistance of the same RSR ternary, and of that ternary with additions of less than 1% Hf and Y and of 3% Cr, and compares this with the oxidation resistance of current Pratt & Whitney F100 engine turbine blades (Alloy 454). The RSR ternary "offers approximately 150°F advantage in temperature capability over our current [DS polycrystalline] metals in the stress range critical for blade design" (15). Research and development work with RSR NiMoAl continues; but neither the results nor an estimate of the possible year of introduction of the material to production manned aircraft gas turbine use is generally available (Proprietary).

2. **Long Range Order (LRO) Materials.** LRO materials are intermetallic compounds. There are three families of these compounds, each with two

members:

Nickel aluminides:	NiAl, Ni_{3Al}
Titanium aluminides:	TiAl, Ti_3Al
Iron aluminides:	FeAl, Fe_{3Al}

Each of these six materials has low density, and is stable, strong, and demonstrates excellent oxidation resistance at elevated temperatures. These properties have been known for many years but none of the materials has seen service in gas turbine engines to date because, in polycrystalline form, each is inherently brittle having low ductility especially at ambient temperatures. This results in poor fabricability, low fracture toughness, and low thermal shock resistance. However, it has been discovered that this low ductility can be overcome by microalloying. This raises the prospect of using the intermetallics in gas turbines. Very little information is generally available concerning these materials (Munitions Act). None of these materials contains any Co or Cr.

a) Nickel aluminides. Ni_{3Al} is the basic γ' phase that imparts high temperature strength to the Ni-based and FeNi-based superalloys. By itself, however, polycrystalline Ni_{3Al} is not structurally useful because of its low ductility. Yet it is well known that single crystals of Ni_{3Al} are ductile. The ductility problems of the polycrystalline form is a grain boundary problem. Research at the Oak Ridge National Laboratory (17) has demonstrated that small additions of boron (around 0.05% B) removes this ductility problem; the resulting B-doped aluminide can be successfully extruded, forged, and cold rolled. In order to preserve weldability, smaller additions around 0.02% B may be used, although this does compromise both fabricability and weldability.

These results are very encouraging and the aircraft gas turbine industry is beginning to evaluate the material. The tensile strength of the Ni aluminides is compared with that of Waspaloy, Hastelloy X, and type 316 stainless steel in Figure 20. Polycrystalline Ni_3Al plus B is being considered for discs (operating temperatures up to 1250C or 2280F seem possible) and its use for combustor liners and turbine vanes is an interesting prospect--the melting point is about 1400C or 2550F. A single crystal Ni_3Al turbine blade is also of interest.

b) Titanium aluminides. The ductility problems of Ti_{3Al} are overcome by

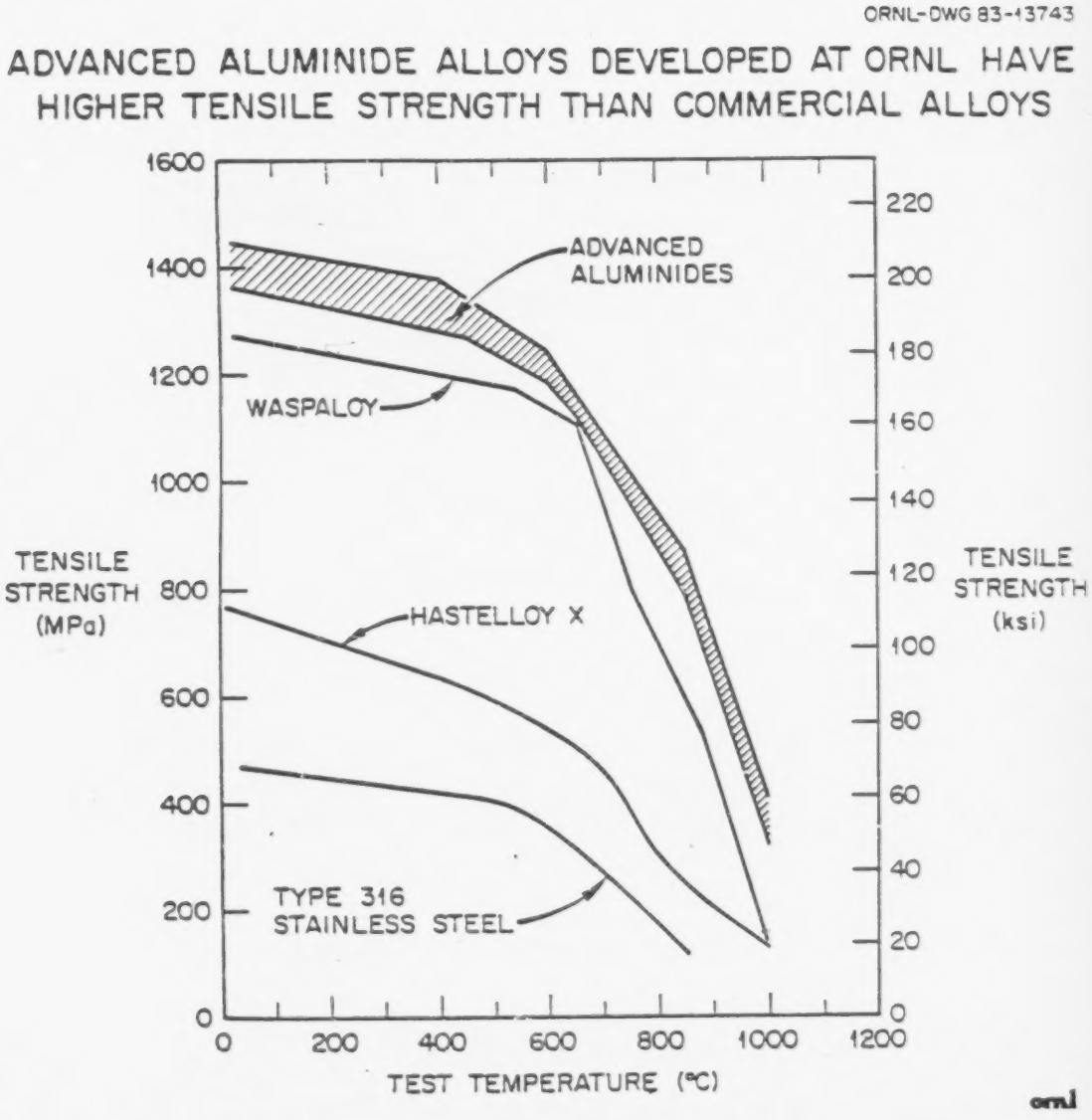


FIG. 20 TENSILE STRENGTH OF ADVANCED NICKEL ALUMINIDES

SOURCE: OAK RIDGE NATIONAL LABORATORY

microalloying with Cb (7). Ti_{3Al} is being studied in the U.S. for use in aircraft gas turbine engine compressor rotors and blades, and low pressure turbine discs, blades, and vanes. The material appears sufficiently strong to be used for blades up to a temperature of 1350F but its oxidation limit is approximately 1200F (7). TiAl appears useful for turbine discs and blades, and in sheet form for cases; its temperature limit for blades is around 1500F and its oxidation limit is approximately 1800F (7). These materials are candidate (albeit expensive) substitutes for "intermediate strength superalloys such as Waspaloy and INCO 713" (6).

c) Iron aluminides. Pratt & Whitney Aircraft has studied these materials in conjunction with RSR rapid solidification (15). The inherent ductility problem was removed and an increase in tensile strength achieved by microalloying with titanium diboride, TiB_2. Their particular interest in these materials is for use as sheet for combustor liners--the materials are useful here up to tempertures of 1800-2000F. The iron aluminides are also of interest for turbine discs, blades, and vanes at lower temperatures, and for cases.

All of these intermetallic compounds are in relatively early stages of development. While they have shown great promise to date, much remains to be learned concerning the full range of their material properties.

3. **Directionally Solidified Eutectic (DSE) Superalloys.** A large number of these materials have been under development during the past two decades, in the U.S. and overseas. They are produced in both polycrystalline and single crystal form. As the crystals are directionally solidified, strong fibers or platelets form in the direction of the crystal growth and substantially improve the material properties in that direction; fibers have been found to be superior to platelets. Two promising groups of eutectics are Ni-based superalloys reinforced with tantalum carbide fibers (NiTaC series) and Co-based superalloys with tantalum carbide fibers (CoTaC series). The development of a turbine blade using the eutectic NiTaC-13 is described in Ref. 18. When used for turbine blades, the fibers are aligned with the blade length to help support the centrifugal load. Figure 17 shows that eutectics are expected to increase the operating temperatures of turbine blades by about 100F above those achieved by single crystal Ni-based superalloy blades. The

major material problems preventing the introduction of DSEs to hot-section service are poor transverse properties and poor oxidation and hot-corrosion resistance. A further impediment is manufacturing cost--this is high because the solidification rates required are considerably lower than present DS and SC blades, and so each blade takes longer to withdraw from the furnace. NiTaC-13 has a furnace withdrawal rate of 1/4 inch per hour compared with a typical Ni-based SC blade withdrawal rate of 10-15 inches per hour. The production of DS eutectic blades in commercial quantities would require many more furnaces than are currently required.

Details of the composition, microstructure, and fabrication processes for DSEs have not been generally released (Munitions Act). These materials typically contain 3-10% Co and 4% Cr.

4. **Oxide Dispersion Strengthened (ODS) Superalloys.** These materials are produced as mechanically alloyed powders that are subsequently hot-extruded and hot-rolled into bars or hot-extruded and then rolled into sheet; bar material may be forged (in fine grain form) or machined into desired shape. The final product is then heat-treated as desired (19). These materials are alloys of Ni or Fe, relying for their high temperature strength on an even dispersion of fine particles of Yttrium oxide, Y_2O_3. One of the Ni-based ODS alloys also contains the strengthening γ' phase for intermediate temperature strength.

NASA has been studying and developing ODS alloys for over two decades, and three ODS alloys are presently commercially available from Inco Alloy Products Company: MA-754, MA-956, and MA-6000. All of these ODS superalloys contain from 15 to 20% Cr and they are all cobalt free. Production capacity for these alloys is presently 120,000 pounds per year and will be escalated to 240,000 pounds in 1984 and 440,000 pounds in 1986 (20).

MA-754 has been used for several years by General Electric for high pressure turbine vanes in a military engine. This Ni-based alloy has a very high burnout temperature, excellent resistance to distortion, and good thermal fatigue resistance. MA-754 vanes have accumulated many hours of service lasting the life of the engine without replacement and no MA-754 vane has ever been removed from a gas turbine engine for cause.

As part of NASA's Materials for Advanced Turbine Engine (MATE) Program, NASA and Pratt & Whitney Aircraft are currently evaluating the Fe-based alloy MA-956 for use in combustor liners. In a special, shingled design devised for this alloy, it appears to have "the potential of extending the operational life of combustor liners to four times their current life" (21). Alternatively, it may be used at a temperature about 300F greater than the common combustor liner material, Hastelloy X. MA-956 is weldable; however, the yttria dispersoid will float to the surface of the weld pool in most fusion welds, thus lowering the elevated temperature strength somewhat. Consequently, structures requiring welding are designed so that welds are placed away from the hottest regions of the structure.

Also under the MATE program, NASA and the Garrett Turbine Engine Company are evaluating the Ni-based alloy MA-6000 for application as a turbine blade alloy (21). Results to date indicate that an increase in metal surface temperature of about 150F over currently used DS blade superalloys may be possible. NASA estimates that ODS superalloy turbine blades have the potential to operate at surface temperatures of 2100F (Fig. 17) provided that sufficiently protective coatings are developed. MATE flight tests of MA-956 and MA-6000 in their respective applications are scheduled for 1984.

Because the mechanical alloying process was invented in 1968, it is still considered to be an emerging technology and accordingly present applications are limited. ODS superalloy tensile strength and intermediate temperature (1400F-1600F) stress-rupture strength are not as great as other superalloys (although their high temperature stress-rupture strength is greater) and this has so far precluded their use for turbine blades. A boon for ODS would be the development of near-net shape capability with ODS powder, but this has not yet occurred.

ODS alloys are generally more expensive than conventional investment cast superalloys but final machined parts cost only about as much as similar DS investment cast parts. As the applictions for these alloys has expanded, part price has been reduced substantially, and further price reductions (as great as 50%) are anticipated as the market grows for turbine and industrial applications. Other fabrication techniques for the high pressure turbine vane application are being examined at present. Prototype forgings have been made

which suggest that overall cost reductions of at least 20% are possible for this part (20).

5. **Fiber-Reinforced Superalloys.** All of the structural materials considered here so far fall into the category of traditional metal alloy systems inasmuch as they are derived from a homogeneous alloy melt. Several other classes of materials exist that have the potential for increasing gas turbine operating temperatures above the maximum possible with these alloy systems (this maximum is around 1150C or 2100F for turbine blades). Among these are a number of composite materials including metal wires in a metal matrix, ceramic fibers in a metal matrix, ceramic fibers in a ceramic matrix, and carbon-carbon composites; in addition to these composites there are monolithic ceramics. Of all these materials, the metal wire/metal matrix composites--called Fiber-Reinforced Superalloys (FRS)--promise the earliest development success although they offer the smallest increase in operating temperature. FRS technology is under development in the NASA Metal Matrix Composite Program. The aim of this program is to develop the technology to fabricate an FRS turbine blade capable of operating at a surface temperature of 1100-1200C or 2000-2200F (22,23).

The problems involved with developing FRS materials are far more complex than those of traditional metal alloys. The fibers and metal matrix play different roles in these composites and therefore require different properties; yet they must be bonded together into a chemically and mechanically stable structure. The separate materials property requirements of the fibers and matrix are as follows (22):

a) Fiber Property Requirements:
   - Stress-rupture and mechanical fatigue strength to enable at least a 50C temperature increase in blade temperature
   - High fracture toughness at high temperatures
   - Low cost

b) Matrix Property Requirements:
   - Chemical compatability with fibers at both fabrication and operating temperature to prevent degrading of fiber strength due to interdiffusion
   - High mechanical and thermal fatigue strength

- Mimimum thermal expansion mismatch between matrix and fiber--such mismatch induces thermal fatigue
- Oxidation and hot-corrosion resistance. The matrix must also shield the fibers from the oxidizing environment
- Low density to offset high density fibers
- High fracture toughness and ductility at low temperatures
- Adequate shear stress-rupture strength

The first generation FRS material selected by NASA for development is a FeCrAlY matrix strengthened by tungsten alloy wire (this tungsten fiber reinforced superalloy is termed TFRS). This composite satisfies the materials property requirements as a **development prototype** and indicates that FRS has the potential to satisfy the turbine blade temperature goal of 1200C. It has been successfully formed into the geometry of an air-cooled turbine blade but such a blade has not yet been engine tested.

> In general, relatively strong conventional nickel and cobalt alloys have been unsatisfactory as matrix alloys for TFRS because of fiber/matrix interdiffusion and property loss. These alloys also are not effective in resisting strain resulting from thermal expansion mismatch. Weak, ductile, oxidation-resistant, iron-base coating alloys (Fe, Cr, Al, Y) have been the best compromise compositions identified to date (23).

The Fe-based TFRS matrix contains no Co and about 15% Cr; the fibers contain no Co or Cr.

There are still some major problems to be overcome before FRS will be acceptable for aircraft turbine blade production.

a) The expansion coefficient of the tungsten fibers is approximately one-third that of the matrix. This induces thermal strain in the blade which can reduce its life considerably unless matrix strain can be absorbed readily--this is the most serious limitation of the technology at present.

b) "Matrix alloy strength is less than optimum to satisfy the requirements of ductility (to absorb thermally induced strain) and interdiffusion resistance. Improved matrix alloy strength without sacrificing interdiffusion performance is needed to permit effective use of higher strength reinforcing wire. Present matrix strength limits the performance gain possible from TFRS" (24).

c) "Oxidation of tungsten fibers is a concern at turbine-operating temperatures. Ductile coatings such as FeCrAlY have shown excellent

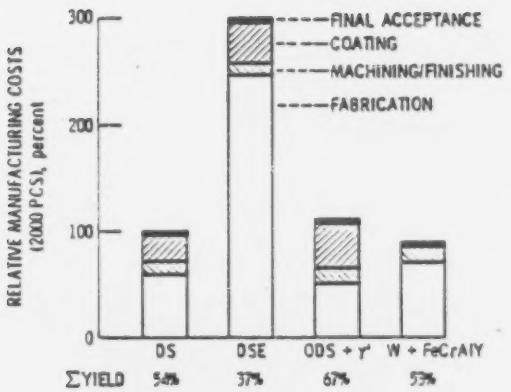


Figure 40. - Cost analysis for JT9D first stage turbine blade. (Ref. 58)

FIG. 21

SOURCE: REF. 22

oxidation and corrosion resistance, however, coating damage or a deep crack would expose fibers to the oxidizing environment. Exposure of fibers by removal of the surface protective coating, such as by foreign object damage, would expose a limited number of fibers for a small decrement in component strength. A more severe threat is possible where a deep crack can expose a significant number of 'fiber ends.' If oxidation of these fibers over their length would occur, the loss could be catastrophic. However, data obtained from a limited number of tests has indicated only a shallow penetration of oxidation along the fiber length from an exposed end. The oxidation products effectively clogged the small hole in the matrix after about a 0.1 inch depth of penetration. More data are necessary to develop confidence that such performance is the norm. Such data to be obtained for conditions simulating the range of service conditions" (24).

Notwithstanding these problems, FRS has great potential as a turbine blade material and may find considerable application to combustor liners, turbine vanes, and other hot-section components. However, current composite design tools are inadequate for predicting FRS failure modes and developing acceptably reliable designs; and current fabrication technology is only at the laboratory stage of development.

The thermal expansion coefficient of TFRS is low compared to superalloys and overlaps that of current thermal barrier coatings (TBCs). Such coatings applied to TFRS components may therefore show less tendency to spall than they do with present superalloys (spalling problems prevent TBCs from being applied to present superalloy blade and vane airfoils). This may constitute a significant advantage for TFRS. Further, regardless of the application of a TBC, the high thermal conductivity of TFRS permits the use of simpler internal blade geometries for cooling air, thereby lowering the manufacturing cost.

Finally, to some cost estimates. Fig. 21 (22) shows estimates of the relative fabrication cost of an air-cooled JT9D-7F engine first stage turbine blade, fabricated from a directionally solidified (DS) superalloy, a directionally solidified eutectic (DSE), an oxide dispersion strengthened superalloy (ODS), and TFRS. The costs of DS, ODS, and TFRS are essentially

the same, and 1/3 that of DSE.

6. **Ceramic-Metal Composites.** Ceramic whiskers and fibers have been used to strengthen a metal matrix (23). Such reinforcement is attractive because ceramic whiskers and fibers have high strength at ambient and elevated temperatures, high elastic modulus, good oxidation resistance, and low density. However, high whisker strength is obtained only with flaw-free whisker surfaces which are difficult to retain due to whisker-matrix interaction. Further, whisker-reinforced composites are difficult to fabricate because small inter-whisker distances are necessary to obtain the high whisker density necessary for composite strength. The high temperature properties so far obtained from whisker-reinforced composites have been poor. Attempts to utilize long single crystal ceramic fibers have also been disappointing because of fiber-matrix interactions and thermal expansion mismatch of fiber and matrix--a combination of these leads to disbonding of fibers or to fiber fracture with attendant loss of composite strength. These problems have not yet been solved.

7. **Monolithic Ceramics.** The term ceramic derives from the Greek word Keramos meaning "burnt stuff" and traditionally refers to the family of earthenware, brick, glass, porcelain, and enamels formed by firing minerals at high temperature (25). Use of the term has now broadened to include the families of materials shown in Table 6. When they contain no internal reinforcement (e.g. by means of fibers, wires, or rods) these materials are called monolithic ceramics.

Engineering interest in monolithic ceramics for structural use in gas turbine engines began in the late 1940s when the importance of their high temperature capability for engine performance was recognized. However, the ceramics available at that time had poor resistance to thermal shock, and little was known about designing with brittle materials. Consequently, early attempts to use ceramics in this application were unsuccessful. The situation changed in the 1960s with the introduction of some new monolithic ceramics--in particular, silicon carbide (SiC) and silicon nitride (Si_3N_4). These materials have greatly improved resistance to thermal shock and much effort has gone into their development (26). Subsequent developments have introduced other members of Table 6 in composite structural materials and as coatings.

Table 6

Families of Ceramic Materials and Their Applications

Ceramic Family	Applications
Glasses	Containers, windows, lenses, insulators, fiber optics, electronics
Silicates, Phosphates	Cements, construction materials, heat exchangers, table ware, sanitary ware, exhaust emissions control, pyroceram^{R} (Corningware^{R}), phosphors
Oxides	Electronics, refractories, abrasives, electromagnetic windows, energy storage, oxygen sensors, bearings, cutting tools, thread guides, extrusion dies, pigments, coatings
Carbides	Abrasives, wear resistance components, seals, heat engine components, heat exchangers, nuclear, heating elements, kiln furniture, coatings
Nitrides	Heat engine components, bearings, electromagnetic windows, cutting tools, heat exchangers
Ferrites and Titanates	Magnetic, electronic, dielectric
Carbons and Graphite	Fixtures, heating elements, seals, rocket nozzles, composite structures

Source: Ref. 14 (D. Richerson)

A number of federal government programs have funded engineering ceramic development since the late 1960s--these have supported basic research and technology development as well as applications projects. Included in the latter are three current DOE-sponsored and NASA-managed projects: Ceramic Applications in Turbine Engines (CATE); Advanced Gas Turbine for Automotive Applications (AGT); and Heavy Duty Transport Technology. "These programs have been successful in applying finite element design methods to ceramic components, developing ceramic component fabrication techniques, and demonstrating that ceramics can operate in heat engines [specifically gas turbines], albeit sporadically...But, the present technology base is inadequate to support successful ceramic applications in heat engines" (27). Nonetheless, monolithic ceramic development for heat engines continues and is extremely encouraging. In parallel with this effort, basic research is ongoing to improve understanding of monolithic ceramics and their processing.

There are several methods for fabricating ceramic parts. Melt formation is used to form glass and single crystal ceramics. Other monolithic ceramics are usually formed by the consolidation of powder. The powder is first formed to the desired shape by various means including isopressing, injection molding, and slip casting. This results in a "green body" preform which is then densified by one of several means. A well bonded polycrystalline monolithic ceramic may be formed by the application of heat to the green body (this process is called sintering) or by the simultaneous application of heat and pressure (two such processes are hot-pressing and hot-isostatic pressing). These processes may be performed in various atmospheres (such as argon or nitrogen). Reaction bonding and reaction sintering are common sintering processes.

For a ceramic to be useful as a structural material in the gas turbine hot-section, and particularly for rotating turbine components, it must exhibit the following properties up to the operating temperature of interest:

- High strength (tensile, shear)
- High thermal shock resistance
- Low creep and good stress-rupture life
- Structural stability over time (low static fatigue)
- Good oxidation and hot-corrosion resistance
- Good erosion resistance

- Adequate fracture toughness
- High reliability (high Weibull modulus)
- Be readily fabricated at reasonable cost

The monolithic ceramics that have been found to best satisfy these requirements to date are silicon carbide and silicon nitride. Turbine inlet temperatures around 2500F are possible with these ceramics, without cooling (Fig. 22). These materials have the additional advantages of low density (which not only reduces engine weight but also reduces the centrifugal loads that rotating components must bear) and of containing no strategic materials. However, these materials still have some significant drawbacks.

a) Reliability is still unacceptably low. "The low reliability results from a combination of the inherent brittleness of the material and the inclusion of very small flaws...in the material that can initiate failure. These flaws are introduced into the material as inclusions, voids, or surface defects during the processing of the material" (28). The low reliability is reflected in the wide scatter of strength property data exhibited by samples of nominally the same material (low Weibull modulus). Improving this reliability is the focus of much ceramic research throughout the country. The problem is being attacked on four fronts: improving the materials themselves; improving processing of the materials; improving the technology of non-destructive evaluation in order to detect smaller flaws; and improving the capability of ceramic component life prediction.

b) The materials are inherently brittle due to their low fracture toughness. This has two consequences, both of which can largely be overcome by design. First, the use of monolithic ceramics for turbine blades is undesirable (particularly in manned vehicles) because of the risk of foreign object damage. This risk can be overcome somewhat by design through the use of reverse flow combustion chambers and other particle separators. Second, the material cannot locally deform plastically in regions of stress concentration (as do metals) and this readily leads to failure. Improvements in design methodology (such as the wider application of finite element analysis) and component proof testing can circumvent this problem in many cases. However, advances in joining technology are also necessary in order to avoid excessive

ADVANCED CERAMICS HAVE SUPERIOR HIGH TEMPERATURE STRENGTH

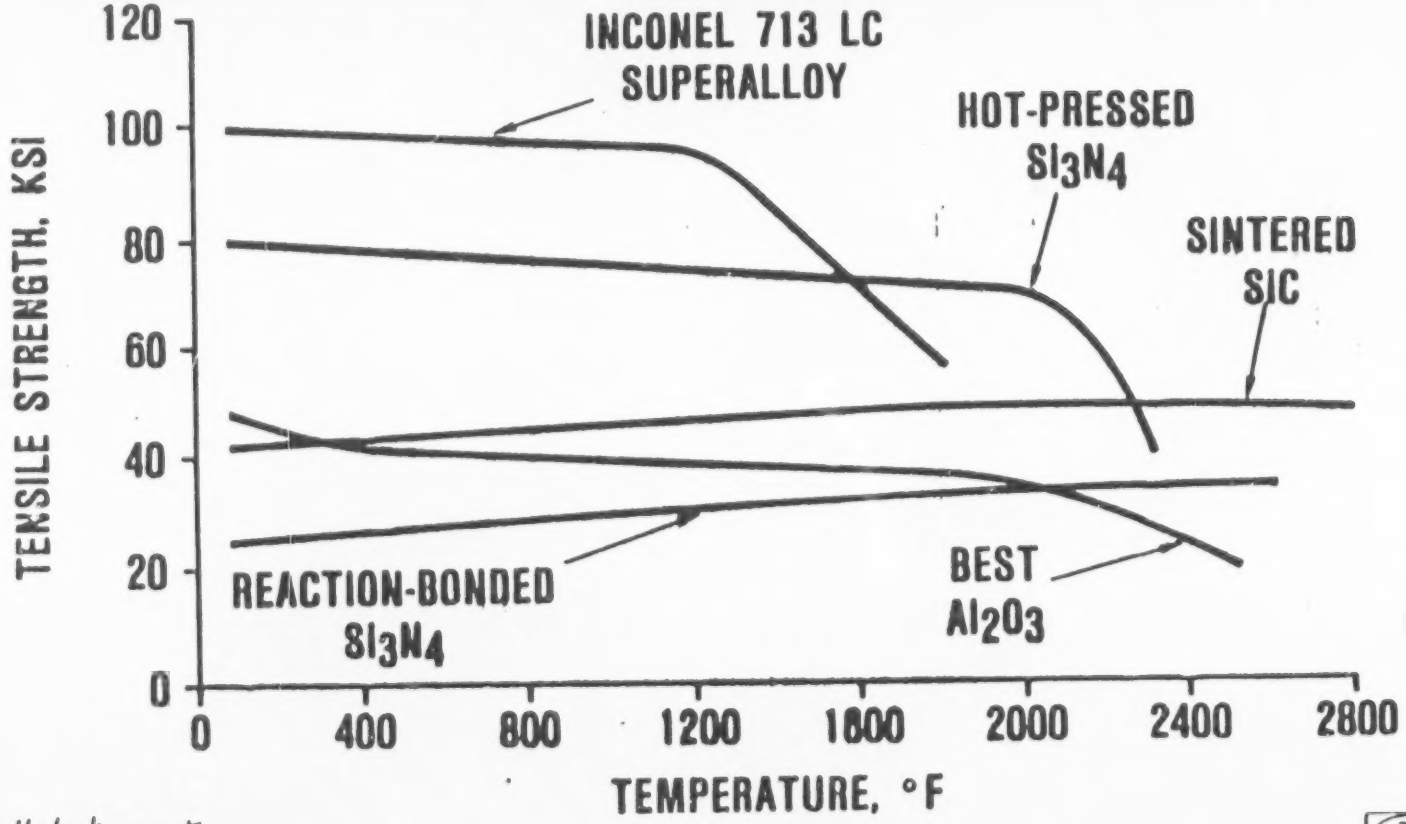


FIG. 22 High temperature strength of monolithic ceramics

Source: Garrett Turbine Engine Company

stress concentrations in and around ceramic-ceramic and ceramic-metal joints.

Despite these problems, monolithic ceramics have been successfully tested for short times in gas turbine engines. For example, a Garrett T76 turboprop engine was redesigned by Garrett "to permit uncooled operation at average turbine inlet temperatures of 1204°C (2200°F)...the redesigned engine contained 102 separate ceramic components [including turbine vanes, blades, and seals and certain support structures]. It was successfully operated in two-hour cycles to full design conditions. A 30% increase in power and a 7% reduction in fuel consumption were demonstrated, compared to the baseline metallic engine" (26). A redesign was conducted which overcame a problem of "contact stress fractures at static structure interfaces" (26). Additional testing of ceramic blades inserted in a superalloy disc in a modified T76 engine achieved the goal of 15 hr of cyclic operation to a turbine inlet temperature of 2200F.

A further promising example of the development of monolithic ceramics for hot-section use is the engine developed by the Allison Gas Turbine Operations of General Motors in the CATE program.

> The baseline engine in the CATE program is the Allison GT404 engine, which, in its all-metal configuration, operates at a turbine-inlet-temperature (TIT) of 1835°F. The TIT was increased to 2070°F by incorporation of ceramic gasifier turbine and regeneration systems. A total of 76 ceramic parts are in the 2070°F configuration gasifier turbine section, that include 28 vanes, three flow-path rings, four flow-path struts, 40 rotor blades, a plenum and a turbine tip shroud [Fig. 23]. With the CATE program concluded early in 1983, a total of 9400 hours of rig and engine testing and 6600 miles of vehicle demonstration have been accumulated on ceramic components, including successful engine demonstration at 100% design speed and 2070°F gasifier rotor inlet temperature (29).

In 1978, Congress passed the Advanced Automotive Research and Development Act. As a result of this, DOE and NASA initiated the AGT program and awarded two contracts to a) Garrett and Ford, and b) General Motors (Allison and Pontiac divisions). Both these contracts involve the design, development, and test of an advanced ceramic automotive gas turbine engine. In the final versions of these engines, ceramic components will not merely be inserted into an existing engine; rather the engines will be designed from scratch to more

Detroit Diesel Allison
Division of General Motors Corporation
GM

CATE Program

2070°F Engine Design Configuration

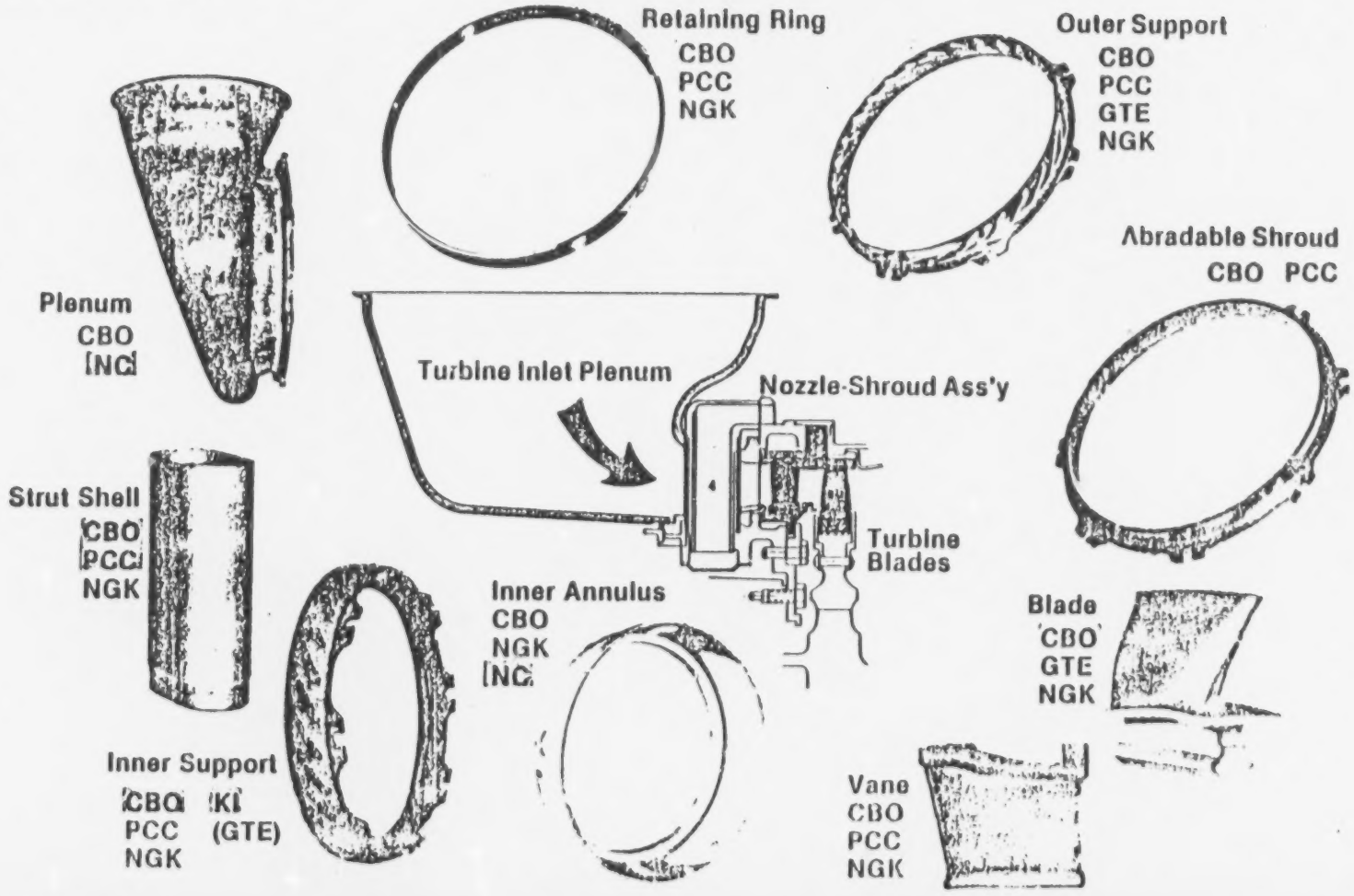


FIG. 23

fully exploit the material properties of the ceramics (26). Work is ongoing in these contracts.

Much work remains to be done to develop monolithic ceramics and their processing, and to gain applications experience before these materials will be acceptable for manned aircraft gas turbine engines. NASA estimates possible introduction of monolithic ceramics as turbine blades in this application around the year 2010 (Fig. 17). During the interim and subsequent periods, hot-section ceramic parts are likely to see service (in static and then rotating parts) in turbosuperchargers and gas turbines progressively as follows:

- Turbosupercharger: static parts and turbine. The turbine material will probably be sintered silicon nitride and will probably appear in the rotating turbine by 1985
- Small stationary electric power generators, similar to the airplane auxiliary power unit (APU)
- Large stationary electric power generators then mobile APUs (ground and marine)
- Short life turbojet and turbofan for missiles: for the cruise missile with ceramic hot-section, NASA predicts a 30% thrust increase and reduced engine weight (30)
- Heavy duty turboshaft vehicular propulsion engine: military truck, tank, off-road vehicle, marine engine
- Automobile propulsion
- Manned aircraft propulsion: NASA predicts a 16% reduction in fuel consumption in commercial transport aircraft due to ceramics (30)
- Utility electric power generation

In order to achieve these advances, continuing development efforts are required in the following areas:

- Improvements in monolithic ceramic materials properties, particularly strength, reliability, and brittle behavior
- Processing and fabrication techniques to give reproducible properties (low scatter) at low cost
- Joining technology
- Non-destructive evaluation and proof-testing

- Design methodology and design data base

The advantages of ceramics with regard to engine performance and zero use of strategic materials are powerful incentives for this massive development task. NASA has developed a plan to advance this important technology and will include this program as a part of its FY85 budget for approval. This plan has been coordinated with the Department of Energy and industry (31).

8. **Ceramic-Ceramic Composites.** As noted above, monolithic ceramics are inherently brittle, exhibiting low fracture toughness. The fracture toughness of these materials can be improved by reinforcing them with ceramic fibers. "The fibers contribute both to increased strain-to-failure of the material and to limited growth of fatal cracks" (28). Unfortunately only a few ceramics are available as fibers; those that are include silicon carbide and carbon, and both have been used in ceramic-ceramic composites. NASA is studying silicon nitride reinforced with silicon carbide fibers, and oxide glasses reinforced with silicon carbide or graphite.

One promising composite is United Technologies' lithium aluminum silicate glass reinforced with silicon carbide fibers. This material has roughly ten times the fracture toughness of the unreinforced glass up to 2000F, but at higher temperatures the glass softens (28).

Reinforcement of higher-temperature ceramics has been relatively unsuccessful to date because of fiber strength losses. These losses are related to:

a) Interaction between fiber and matrix at the higher fabrication temperatures
b) Fiber recrystallization
c) Fiber fracture due to abrasion and stress during processing

As a result of these problems, NASA research has been redirected to ceramic composites employing weak bonding between the fiber and the matrix (23) and to developing processing technologies that employ lower temperatures and no pressing (28). Research is also being directed at improving design tools for optimizing the strength and toughness of these composites. Clearly, ceramic-ceramic composites are still in their infancy; they may eventually be used for combustor liners and for turbine vanes, blades, and cases.

STRENGTH OF C-C COMPOSITES

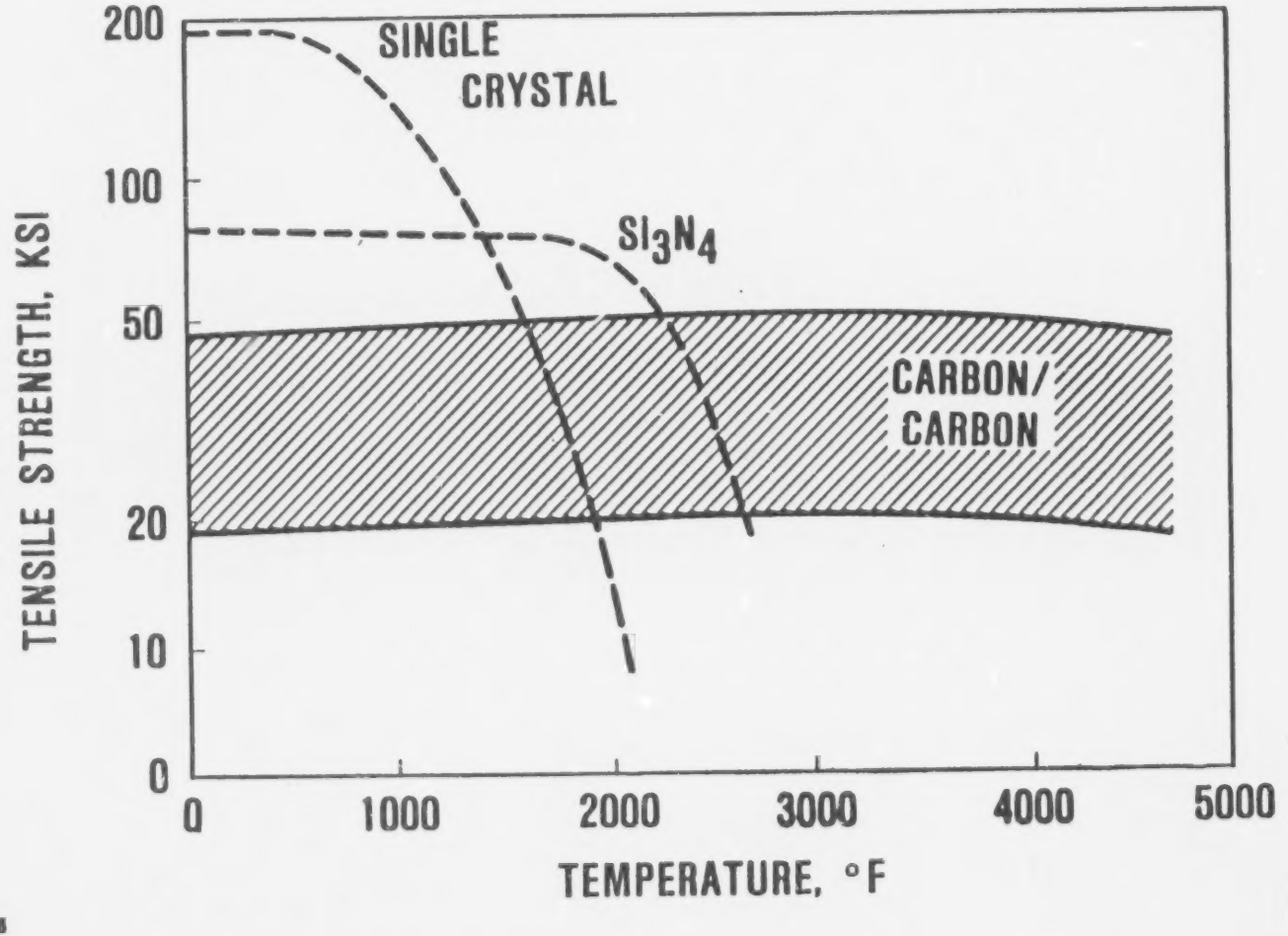


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FIG. 24 Strength of carbon-carbon composites
Source: Garrett Turbine Engine Company

9. **Carbon-Carbon Composites.** These composite materials belong to the family of ceramic-ceramic composites, but they are usually treated separately. They comprise a carbon matrix reinforced with carbon or graphite fibers, are very strong and tough, and have the potential for maintaining these properties at temperatures up to 4500F (28)--see Fig. 24. If this doesn't surprise you at first, it should. When was the last time you burned coal or wood? Of course these composites can only be used at high temperatures when they are completely isolated from oxygen; in the gas turbine engine this isolation must be provided by a virtually infallible coating. In addition to promising high temperature service, these composites are very light--somewhat less than 25% superalloy density and about 50% ceramic density (monolithic or ceramic-ceramic composite)--are dimensionally stable, have low coefficients of thermal expansion, require no cooling, and contain no strategic materials (32).

Today, carbon-carbon composites can be formed as solid sheet, ribbed sheet, sandwich sheet (carbon felt faced with carbon-carbon sheet), and as truss-core panels. In these forms the composite may be useful in the gas turbine for combustor liners, cases, afterburners, and nozzles. The composite has also been formed into a one piece bladed turbine disc (28). A typical coating for carbon-carbon composites is silicon carbide.

Development of carbon-carbon composites at NASA and in industry is aimed at reducing fabrication costs (which are at present very high), improving strength and stiffness, improving the reliability of coatings, and incorporating oxidation inhibitors into the carbon matrix to prevent catastrophic oxidation in the event of coating failure. NASA predicts that carbon-carbon composites may be introduced as manned aircraft turbine blade materials around the year 2010 (Fig. 12). Prior to that usage, these composites will see action in the afterburners and nozzles of manned aircraft engines, and wider use in unmanned aircraft engines.

THE ROLES OF Co AND Cr IN HOT-SECTION MATERIALS

In order to rationally evaluate the potential of alloy substitution for Co and Cr in hot-section materials--specifically the superalloys--we must

first determine what roles Co and Cr play in enhancing or degrading the properties of those materials. These roles are a matter of some controversy amongst metallurgists even though considerable experimental work has been performed to determine them. One reason for the controversy is that any given superalloy is an extremely complex metallurgical system in which the synergistic effects of the many alloying elements are important. Another is that the roles of Co and Cr within a given alloy are often (but certainly not always) subtle. Further, we compound the problem of defining roles when we seek to generalize roles determined in a few specific alloys to the whole family of superalloys—we try to generalize specific results because it is simply too expensive to obtain data for every alloy used.

Co and Cr affect the mechanical properties of superalloys inasmuch as they affect alloy microstructure. Metallurgists do not all agree on how Co and Cr affect microstructure; nor is it always clear how microstructure and a given mechanical property are related (33).* There are insufficient experimental data to establish these relationships.

However, the experimental work that has been done is valuable. A sample of this work is reported in Refs. 34-36; this work was all stimulated by the recent cobalt shortage and is all directed at the role of cobalt and the potential for decreasing or eliminating its use in superalloys. The work concerned the following alloys.

- Maurer et al (34) — Waspaloy (moderate strength wrought disc alloy)
- Nathal & Maier (35) — MAR-M247 (high strength cast blade and vane alloy)
- Tien & Jarrett (36) — Udimet 700 (essentially Astroloy and Rene-77; used in course grained cast and wrought form for blades, and in fine grained wrought form for discs)

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* The remarks of J.R. Stephens (33) are specifically directed to cobalt and its role in Ni-based superalloys.

As a result of the experimental work done to date, metallurgists have reached general agreement concerning most of the gross effects of Co and Cr on superalloy properties despite the fact that some microstructure questions remain unresolved. What follows is an attempt to report that consensus, and was gleaned from the literature as well as from discussions with industry and academia (14); no attempt is made to summarize specific experimental work.

The approach taken here is first to discuss the roles of Co and Cr in conventionally cast and wrought superalloys; the Ni- and FeNi-based alloys are treated together and the Co-based alloys are treated as a second group. Variations in these roles for each of the other current superalloy macrostructures (DS and single crystal), and for each of the hot-section materials under development which were discussed in the foregoing section, are then addressed. Some brief remarks treat coatings.

Conventionally Cast and Wrought Superalloys

The roles of Co and Cr in conventionally cast and wrought Ni-based and FeNi-based superalloys are summarized in Table 7; and those of Co and Cr in the Co-based superalloys in Table 8. In these tables the various properties are separated into three groups: mechanical (top), physical, and corrosion (bottom). Further, the roles of Co and Cr are presented qualitatively in terms of material property **sensitivity** to Co and Cr content. To explain the differences between these tables we need to know how the three superalloy families differ from one another.

The major difference between the superalloy families is the mechanism by which they obtain their strength. Each of the three families is strengthened somewhat by the solution of the various alloying elements in the basic Ni, FeNi, and Co matrix (respectively) of the metal grains--this is called solid solution strengthening. Co and Cr provide only minor solid solution strengthening of the Ni- and FeNi-based alloys; and Cr only minor solid solution strengthening of the Co-based alloys. Further strengthening of each family is derived from small particles precipitated within the matrix of the metal grains and/or along the grain boundaries--this is called precipitation strengthening. The superalloy families differ with regard to the particles

Table 7
Roles of Co and Cr in Conventionally Cast and Wrought
Ni-Based and FeNi-based Superalloys

Material Property	Role of Co or Cr in Determining Material Property: Property Sensitivity to Co Content		Property Sensitivity to Cr Content	
Strength (yield & ultimate)	small +		—	
Ductility (formability)	moderate +	(1)	small +	
Creep resistance	small +	(2)	small +	
Stress-rupture life	small +	(2)	small +	
Fracture toughness	?	(3)	?	(3)
High-cycle fatigue strength	—		—	
Low-cycle fatigue strength	(small +)	(4)	(small -)	(5)
Erosion resistance	—		—	
Wear resistance	small +		small +	
Machinability	—		small -	
Weldability	small +		small -	
Elastic modulus	—		—	
Density	—		—	
Coefficient of thermal expansion	—		moderate -	
Thermal conductivity	small +		?	
Melting point	small +		—	
Oxidation resistance	small - and (small +)	(6)	large +	
Aqueous corrosion resistance	(small +)	(6)	large +	
Hot-corrosion resistance	(small +)	(6)	large +	
Burnout temperature	(small +)	(6)	large +	

Key: + desirable effect
- undesirable effect
? unknown effect
() indirect effect
— negligible or no effect

Notes: (1) Moderate + refers to all sheet and to those few disc materials that have a low content of the strengthening γ' phase. For high γ' materials (turbine vanes, blades, moderate and high strength disc materials) the role is negligible.
(2) Small + refers to materials with very small grains. The "small" grain sizes in todays discs and cases usually exceed the grain size for which the role is important, and so the role is negligible. For large grain sizes used in combustor liners, vanes, and blades the effect is negligible.
(3) No data available, but the role is probably negligible.
(4) Indirect role via effect on thermal conductivity.
(5) Indirect role via effect on thermal expansion.
(6) Indirect role: Co is a host for Cr. In addition to this, Co has a direct small negative effect on oxidation resistance.

Source: OTA

Table 8
Roles of Co and Cr in Conventionally Cast and Wrought Co-Based Superalloys

Material Property	Role of Co or Cr in Determining Material Property	
	Property Sensitivity to Co Content	Property Sensitivity to Cr Content
Strength (yield & ultimate)	small +	moderate +
Ductility (formability)	—	—
Creep resistance	—	moderate +
Stress-rupture life	—	moderate +
Fracture toughness	? (1)	? (1)
High-cycle fatigue strength	—	small -
Low-cycle fatigue strength	(moderate +) (2)	(small -) (3)
Erosion resistance	—	—
Wear resistance	moderate +	moderate +
Machinability	small -	small -
Weldability	large +	small -
Elastic modulus	—	—
Density	—	—
Coefficient of thermal expansion	—	large -
Thermal conductivity	moderate +	?
Melting point	small +	—
Oxidation resistance	(small +) (4)	large +
Aqueous corrosion resistance	(small +) (4)	large +
Hot-corrosion resistance	(small +) (4)	large +
Burnout temperature	(small +) (4)	large +

Key:
+ desirable effect
- undesirable effect
? unknown effect
() indirect effect
— negligible or no effect

Notes:

(1) No data available, but the role is probably negligible.
(2) Indirect role via effect on thermal conductivity.
(3) Indirect role via effect on thermal expansion.
(4) Indirect role: Co is a host for Cr.

Source: OTA

that are precipitated, and therein lies the major difference in their strengthening mechanisms.

The Ni-based and FeNi-based alloys obtain their strength principally from small precipitated particles of the γ' phase which is basically $Ni_3(Al, Ti)$ plus the refractory metals (Cb, Mo, Ta, and W) plus a small fraction (about one tenth) of the total Co in the material. The only exceptions to this statement are the FeNi-based alloys that are wrought into sheet (such as the combustor liner material Hastelloy X)--these contain almost no γ'. A small additional strengthening of the Ni-based and FeNi-based alloys derives from carbides of Cr, Ti, and the refractory metals precipitated mostly in the grain boundaries. These carbides strengthen grain boundaries, resisting the sliding of one grain relative to another. The Co-based alloys contain no γ'; they are precipitation strengthened by the same carbides, both within the grains and in the grain boundaries. Considering the various precipitated particles--their strengthening effects and the alloying elements in them--and the fact that Co and Cr yield only minor solid solution strengthening, would expect that Co and Cr would play a small role in the mechanical properties of the Ni- and FeNi-based families; and that Cr would dominate Co in their effects on the mechanical properties of the Co-based family. This is seen to be the case in Tables 7 and 8.

The role of Co throughout Table 7 is small. That of Cr is small except for its moderate effect on increasing the coefficient of thermal expansion (which tends to exacerbate thermal fatigue) and its major role in fighting oxidation and corrosion and increasing burnout temperature.

In Table 8 the dominant role of Co is to greatly improve weldability; it also promotes wear resistance and increases thermal conductivity (which diminishes thermal fatigue). The role of Cr in mechanical properties reflects its effects on carbide formation; Cr also increases the coefficient of thermal expansion which promotes thermal fatigue. The major role of Cr is to fight oxidation and corrosion and to increase burnout temperature. As Cr content increases in these alloys, oxidation and corrosion resistance increase but weldability falls; the Cr level of a given Co-based alloy is often chosen in a tradeoff between these effects.

It is worth noting here that a "small" sensitivity in Tables 7 and 8 does not mean that the effect of Co or Cr in a given alloy is negligible. First, the actual effect is the product of alloy composition and property sensitivity, so that even a moderate composition level can yield significant property effects even when the property sensitivity is "small." Of course a small composition level together with a small property sensitivity will yield a negligible property effect. Second, hot-section materials operate in a severe environment and are designed and used to the limit of their capability, and small property contributions are important.

Other Hot-Section Materials

For directionally solidified and single crystal macrostructures we need only consider Ni-based superalloys. The roles of Co and Cr in DS alloys are the same as in Table 7. In the case of single crystals there are no grain boundaries and therefore carbides are not needed to strengthen these. The carbon level is usually reduced in single crystal alloys, so that any roles of Co and Cr in Table 7 that are associated with carbides will be different for single crystals--the role of Cr only is affected. The effects of Cr on ductility, creep resistance, stress-rupture life, wear resistance, machinability, and weldability become negligible; the remainder of Table 7 applies to single crystals.

The roles of Co and Cr in rapid solidification processed superalloys and in DS eutectics that contain these elements are the same as in Tables 7 and 8.

ODS superalloys to date contain no Co, and derive their strength primarily from yttrium oxide particles. Any role of Cr associated with carbides will be masked by the oxide particle effects. The role of Cr is therefore to enhance oxidation and corrosion resistance and burnout temperature; it will contribute to thermal expansion and will thereby slightly increase thermal fatigue.

Fiber-reinforced superalloys also contain no Co to date. The role of Cr is the same as for ODS. The remainder of the structural hot-section materials previously discussed contain no Co or Cr.

The role of Cr in coatings is always to enhance oxidation and corrosion resistance (coatings usually contain about 20% Cr for this purpose), although the majority of the oxidation resistance usually derives from Al. The role of Co in MCrAlY coatings and in thermal barrier bond coats is to improve coating ductility, thereby enhancing coating integrity by diminishing any tendency of the coating to spall from the substrate. Co is also a host for Cr, so that more Cr can be loaded into the coating as Co content is increased. In hardface coatings, Co is often the matrix that holds the hard carbide particles; it also provides the weldability that enables such coatings to be applied by welding; the Co oxide that forms on the surface provides good surface lubricity (12).

Now let us return to Table 4 and review the composition levels of Co and Cr in the typical alloys shown there. A number of questions need to be addressed.

1. **Why do the two common combustor liner materials differ in Co content by nearly 40%?** Compared with Hastelloy X, the Co-based HA-188 has a higher melting point, higher burnout temperature, higher oxidation resistance, and a lower tendency to creep-buckling (these advantages are largely due to the fact that HA-188 is Co-based and has no Fe and higher refractory metal content than Hastelloy X), and it has superior thermal shock resistance. HA-188 is used in the highest temperature applications, especially where the design is creep-limited. Unfortunately, HA-188 is nearly twice as expensive as Hastelloy X.

2. **Why do some turbine vane alloys contain over 50% Co while others are Ni-based alloys?** Of all the conventionally cast vane alloys, the two Co-based alloys are used in the highest temperature applications--first stage vanes. The high Co level gives high melting point (and burnout temperature), high thermal shock resistance, and allows the alloys to be loaded with Cr for excellent oxidation and hot-corrosion resistance. Further, these Co-based alloys are weldable whereas the Ni-based alloys are generally not; vane refurbishability is a valuable property to the customer, for it saves money on spares. Ni-based alloys are generally used in the lower temperature turbine stages. When coated they can be used successfully for first stage vanes but cannot usually be weld-repaired. A manufacturer who used Ni-based vanes where the refurbishability of Co-based vanes is a customer preference would be at a

competitive disadvantage, unless the vanes last for the life of the engine as in the case of MA-754. The Ni-based cast DS MAR-M200 is used to 1900F; its directional structure enhances its thermal shock resistance but it must be coated. There are now processes in production for repairing DS MAR-M200 blades and vanes through selective heat-treatment and weld repair; the development of similar processes for other Ni-based alloys is being pursued.

3. **Why is Co content so high in Ni-based disc alloys and generally lower in blade alloys?** Turbine blades operate at very high temperatures and require large amounts of γ' phase for strength--as much as 65% γ' by volume. Almost no Co resides in the γ' phase, so it resides in solid solution in the somewhat restricted matrix region surrounding the γ' particles. The amount of Co in the alloy must be kept low or it will precipitate out of the surrounding matrix and form undesirable phases that weaken the alloy. Discs don't need as much γ' ; therefore Co content can be higher than in blades and is higher in most cases shown in Table 4.

4. **Why do turbine blades (which are immersed in the hot oxidizing and corrosive gases) have so little Cr while discs (which operate at lower temperatures and are only in contact with the hot gases at their rims) have so much Cr?** Turbine blades require large amounts of γ' phase for strength and Cr, like Co, resides mostly in the matrix surrounding the γ' phase. Thus Cr level must be kept low in blades or it will precipitate out of the matrix to form undesirable phases that weaken the alloy. The blades are coated to enhance their oxidation and hot-corrosion resistance. Discs, which do not need as much γ', can contain more Cr and generally do.

5. **Are the Co and Cr levels currently used in the various hot-section superalloys really necessary? Could they be lowered or eliminated completely through alloy modification without degrading the material properties?** These questions are addressed in the following section.

SUBSTITUTION FOR Co AND Cr

There are four basic ways in which we can substitute for Co and Cr in hot-section parts: alloy substitution; the use of coatings; substitution of the various high temperature structural materials under development; and substitution by design.

Alloy Substitution

Four strategies for alloy substitution may be identified.

1. **Alloy replacement:** substitute an existing or new alloy with lower Co and/or Cr content. In many cases this will result either in decreased component life or in the necessity of derating engine performance--e.g., substitution of Hastelloy X for HA-188 in combustor liners; IN-713 for X-40 vanes; or IN-901 for Astroloy discs. Decreasing component life may result in an increase in Co and Cr usage over the life of the engine because of increased component replacement rate. While such substitution of materials with less than desirable properties is not attractive for long-term use, it may be quite acceptable as a short-term solution during a materials supply crisis.

In other cases decreased use of Co and Cr as well as improved performance are obtained by substituting an advanced alloy, but at greatly increased cost --e.g., substitution of Alloy 454 single crystal blades for Rene-80; or Rene-95 for Waspaloy discs. However, increasing use of advanced alloys is a design trend for performance reasons.

2. **Alloy modification.** The idea here is to reduce or eliminate Co and/or Cr from the alloy, maintaining constant material properties through composition modification of the other alloying elements and minor changes in heat-treatment. This method has only limited usefulness for Cr substitution (in the absence of greater reliance on coatings) because Cr is unique in its ability to confer corrosion resistance, and is generally required at a minimum of 10% Cr or at the maximum level the alloy can tolerate below this level. In general, the strength functions of Cr may be substituted by the proposed method but not the oxidation and corrosion resistance functions.

Lowering or eliminating Co content in Ni-based alloys (the greatest Co users)* is feasible. It has been demonstrated that Co content can be significantly reduced or eliminated in some alloys without seriously degrading

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* The majority of Co is consumed by the high volume sales Ni-based alloys. Of these, Waspaloy is the dominant user.

certain material properties (36). It has been suggested that Co can probably be backed out of most Ni-based alloys while the material properties are kept constant through alloying changes in Al, Ti, and the refractory metals, and through heat-treatment modification. This suggestion is based on laboratory test results (36). However, some in industry point out that while this may be true in the laboratory, it would likely prove impractical in a production environment. The reason given is that the presence of Co widens the usable heat-treatment window; i.e., Co increases the tolerance of material properties to heat-treatment temperature and time at temperature. Eliminating Co and varying other alloying element compositions would so shrink the heat-treatment window for constant material properties that current heat-treatment equipment would be incapable of producing alloys with the required (and currently obtainable) low scatter in properties. The heat-treatment window would then have to be widened by further adjustments in Al, Ti, and the refractory metals and this would incur a loss of material properties. This argument requires experimental confirmation and appears to be contradicted by the previously mentioned laboratory results (36); its validity may well depend upon the particular alloy in question.

Alloy modification with constant material properties may be pursued by reducing Co and Cr compositions and adjusting other alloying element compositions only insofar as heat-treatment tolerances will allow. This will be no simple task, for so many material properties are involved--witness Table 7. However, if it proves possible it may allow reductions in Co levels in turbine blade alloys and Ni-based vane alloys to the range 5-10% Co (cf. the new Alloy 454) and in Ni-based turbine discs and cases to the same range (cf. the new alloy Rene-95)—these would represent substantial Co savings. It is impractical to seek substantial Co reduction in Co-based alloys. Cr savings would probably be small (in the absence of greater reliance on coatings) except in discs where it may be possible to reduce Cr composition to 10% Cr, this being the minimum amount necessary for hot-corrosion resistance at the rim.

However, the cost and effort involved in developing a range of reduced-Co Ni-based alloys to replace all the current major sales volume Ni-based alloys is absolutely prohibitive for industry. Costly alloy development, scale-up and qualifiction are involved. For each alloy, this process may cost in the

vicinity of $2-10 million and take 3-10 years depending upon the application (14).

In light of these realities, alloy modification to reduce Co and Cr usage is being pursued by industry only when a new alloy is required for performance reasons. Here usage can be minimized in the normal course of events.

3. **Local Property Tailoring.** When a component is made from a single material, the material must be capable of meeting the most demanding property requirements throughout the component. This often means that the material is chosen because one or more of its properties is required at specific locations. If high Co and Cr content are required in only one region because of temperature extremes (say, in the airfoil region of a turbine vane but not in the attachment region) then the whole component has high Co and Cr content, even where these elements are not needed in such large quantities. In recognition of this, the idea of using two or more materials in the fabrication of hot-section parts was suggested many years ago.* The major problem with implementing this idea is that it requires reliable joints between the various pieces used--and those joints must possess the properties required of the pieces themselves. This requires the development of sophisticated joining technology.

One well recognized application of local property tailoring is the dual-alloy turbine disc. The differences in material property requirements between the hub and rim of the disc (see Table 2) suggests the use of two different materials in its construction--one for the hub and another for the rim. The disc is made of two pieces, a hub and an annular rim, joined together. This allows separate optimization of material in the two regions. Further, the dual-alloy concept might be employed to virtually eliminate Co and Cr from the hub region of the disc--the majority of the disc volume--when the design allowed, thereby introducing substantial Co and Cr savings. However, the potential for this technology is dependent upon the particular application and

*The Germans pioneered this idea (before the end of World War II) in their attempts to construct turbine blades "which were metal at the root, where stresses are highest, and ceramic towards the tip where temperatures are highest" (2).

it has been little used to date for either design optimization or strategic material reduction. The Garrett Turbine Engine Company has successfully used such dual-alloy construction in the turbine of their GTCP331 auxiliary power unit. A description of this application and of its technology development is given in the Appendix (37).

Allison is also developing dual-alloy technology. There are two applications of this technology at Allison. The first is a dual-alloy turbine disc with air-cooled turbine blades--this has been successfully run in the Army ATDE development engine. The second is a dual-alloy turbine airfoil comprising single crystal leading and trailing edges joined to a lower performance alloy center spar--this application is in early stages of development (29).

4. **Grading of Metal Powder.** Powder metallurgy allows the extension of the tailored property idea from a stepwise variation in properties to a continuous one. In principle, the metal powder used to form the component can be graded--i.e., the alloy content of the powder can be varied continuously--throughout the component to achieve optimum property variation or minimum use of strategic materials.

Early development work on this technology performed by NASA and TRW has resulted in the successful fabrication of discs by the consolidation of two different metal powders. The feasibility of dual-powder technology, on the laboratory scale, has been proven at TRW (38) showing that material properties can indeed be varied in this manner; components up to 350 lb in weight have been fabricated. On-going work at NASA-Lewis Research Center is investigating the adjacent use of various metal powders. This work is directed at saving strategic materials and is being performed as part of NASA's Conservation of Strategic Aerospace Materials (COSAM) program. To date, multi-powder components have undergone mechanical property characterization but have not been spin pit tested or engine tested (39).

A number of engine manufacturers are also pursuing this multi-powder technology. However, the work so far has concentrated on the juxtaposition of different powders. Only exploratory work has been done on the continuous grading of powder (39); it is not yet clear that it is possible to design and fabricate components in this fashion.

Coatings

Coatings may be employed in two ways to save strategic materials.

1. **Greater reliance on coating protection.** Metallic coatings are currently employed to enhance oxidation and corrosion resistance (diffusion coatings, MCrAlY overlay coatings, and thermal barrier bond coats) and to increase surface wear resistance. Clearly, Co and Cr content should be minimized in these coatings commensurate with the coating function--this will not provide significant Co and Cr savings except perhaps in wear resistant coatings where attempts are being made to replace Co matrices with Ni (12). However, the use of oxidation and corrosion resistant coatings offers the possibility of considerable Cr (and some Co) savings in the substrate material. Wider use of wear resistant coatings offer some small Co savings through reductions in Co content in substrate materials (12) and/or extension of component life.

The present design philosophy for the use of oxidation and corrosion resistant coatings is very conservative. The substrate material is loaded with as much Co and Cr as is necessary (or possible) in order to provide the resistance. A coating is then applied to provide additional protection and extend component life. Loading of the substrate is necessary because current coatings are not nearly 100% reliable and the manufacturer and operator both wish to avoid possible catastrophic failure of a component should its coating be penetrated. Such penetration can occur during impact with a foreign object, or simply as the result of coating oxidation, corrosion, erosion, or wear.

If oxidation and corrosion resistant coatings approached complete reliability, then it may be possible to abandon the current conservative design philosophy. In that event large Cr savings would be possible in all three families of superalloy substrate materials in most hot-section parts; and some small Co savings in Ni-based alloys may accrue. (Some small Cr content may still be required in the substrate to avoid diffusion degradation of the coating and premature spalling.) Similarly, thermal barrier coatings may allow the use of substrate alloys with lower Co and Cr levels and having lower temperature capability. However, at present there is extreme reluctance

amongst both manufacturers and operators to pursue such a course. It is not clear that coating reliability will ever reach the levels required to permit this.

2. **Life extension due to thermal barriers.** Thermal barrier coatings (TBCs) provide a substantial temperature differential across their ceramic surface layer--a figure of 100-150F is possible today. The designer uses this differential to either (a) increase the turbine inlet temperature while keeping substrate temperature constant, or (b) increase component life at constant gas temperature, or (c) a mix of these. It is remarkable, but a 100F decrease in metal substrate temperature typically increases the life of a superalloy turbine blade by a factor of ten. Increases in component life translate to decreased strategic material usage over the life of the engine due to reduced component replacement rate. Wider use of TBCs therefore has the potential to save considerable strategic material--TBCs may eventually be used on turbine blade and vane airfoils, as well as in their present application to combustor liners and vane platforms. However, manufacturers will take advantage of TBCs for performance increases, (a) above.

Substitution of Structural Materials Under Development

The various structural hot-section materials previously described may be used, if their development is successful, in two ways: for the fabrication of complete components; and for the fabrication of pieces of components for use in local property tailoring.

1. **Fabrication of complete components.** The possible future uses of the various materials have been discussed previously and are summarized in Table 9. The strategic element content of the majority of these materials is negligible or zero. However, the only materials likely to be available in the next 5 years for production use are **rapid solidification processed superalloys** and **ODS superalloys** (ODS are currently in production for turbine vanes); it is uncertain what component applications these materials may have by that time.

Table 9
Use of Structural Materials Under Development to Reduce
Co and Cr Usage in Hot-Section Parts

Material	Composition (% Weight) Co	Cr	Suitable Components
Rapid solidification processed superalloys	0+	0+	Combustor liners; cases; turbine blades, vanes, discs, seals.
LRO (intermetallics)			
Ni	—	—	HP turbine discs; (combustor liners; HP turbine vanes, blades).
Ti	—	—	LP turbine discs, blades, vanes; cases.
Fe	—	—	Combustor liners; LP turbine discs, blades, vanes; cases.
DS eutectics	3-10	4	Turbine blades, vanes
ODS superalloys	—	15-20	Turbine vanes; (combustor liners, turbine blades).
FRS	—	15 (matrix)	Turbine blades, vanes; combustor liners; cases.
Monolithic ceramics	—	—	Turbine discs, vanes, blades, seals; cases; combustor liners.
Ceramic-ceramic composites	—	—	Turbine vanes, blades, cases.
Carbon-carbon composites	—	—	Combustor liners; cases; afterburners; nozzles; turbine discs, vanes, blades.

Key: HP = high pressure
LP = low pressure
( ) = less likely application

Source: OTA

Concerning the RSR process of Pratt & Whitney Aircraft, Moore (40) says:

> Rapid Solidification Rate (RSR) technology is in development and test in the areas of superalloys, aluminum, and steel. Test parts are being fabricated and will be engine tested. It is anticipated that parts could be qualified for use in next generation engines or for substitution in current engines. Limited production capacity for powder exists today and could be expanded easily. Current manufacturing methods for consolidation, forging, machining, and joining can be utilized for production. These alloys could be used for all hot section parts.

Figure 17 has shown NASA's estimates of first use dates for some of these materials in manned aircraft turbine blade applications. By those estimates, DS eutectics should be available now but there is no public indication that they are about to be unleashed. Possible dates for the use of the various materials in applications less demanding than turbine blades are not known.

Of course, the use of powder metallurgy, with or without rapid solidification, in a net or near-net shape manufacturing process will reduce strategic material consumption by eliminating scrap losses in forging and machining of the same alloy. Material weight savings of about 1/2 and cost savings of about 1/3 relative to conventional processing are possible (11). Greater use of powder metallurgy than it currently enjoys is in part dependent upon further developments of non-destructive evaluation techniques for identifying very small internal material flaws. If not detected, these flaws can cause early failure of components in service. Large structural castings of superalloys and titanium are just getting into production and are significantly reducing input material. To date, near-net shape forging has reduced material input requirements, but more can be done. Scrap revert programs have been partially implemented and have been totally successful (40).

2. **Local property tailoring.** The use of monolithic ceramics in local hot-spots has been suggested; and it is possible that these and other structural materials under development may find such usage. However, the joining technology previously mentioned must be developed. No serious moves appear to be afoot to implement local property tailoring using any of these materials.

Substitution by Design

There are three ways in which substitution by design might be implemented to save strategic materials.

1. **Performance downgrading.** Substantial strategic materials savings are possible by designing components to operate at temperatures and stresses somewhat below the maximum possible with given materials. This would extend component life and therefore decrease the use of strategic materials over the life of the engine by reducing component replacement rates. To give an idea of the power of this approach: a 25F decrease in turbine blade metal surface temperature typically increases the blade life by up to a factor of two. In a large commercial transport engine, a 25F drop in turbine inlet temperature would typically decrease thrust by 2-3% and slightly increase trip fuel consumption; this would be accompanied by a significant decrease in maintenance costs due to reduced blade replacement rate (14). A similar approach is the operational derating of engines in the field, in the event of a materials supply shortage.

The higher the temperature capability of hot-section materials, the higher the temperature that could be retained after the temperature drop was exercised. The high temperature materials under development therefore offer the greatest potential for a design performance/life tradeoff. However, there is little reason (from the point of view of saving strategic materials) to exercise this tradeoff with most of these materials since they contain very little of the strategic elements.

The substitution methodology of performance downgrading is in direct opposition to past and present design practice. It is extremely unlikely that it would ever be implemented unless the industry were faced with an extreme strategic materials supply situation, or a government mandate prevailed.

2. **Increased component efficiency.** The aerodynamic and thermodynamic efficiencies of compressor and turbine blades and their associated vanes dictate in part the total number of compressor and turbine stages required for a given engine performance. These efficiencies are continuously improving due to advances in both theory and practice. This raises the prospect of reducing the number and/or size of certain parts, thereby saving material.

The trend of increasing component efficiency has reached the point where the next generation of aircraft gas turbine engines (beginning with military airplane engines) will employ fewer stages in their compressors and turbines. The number of compressor rotors, stators, and blades, and the number of turbine discs, vanes, and blades will thereby be significantly reduced. This will reduce the quantity of strategic materials used in the initial engine build, and also that used annually for spares--assuming replacement rates do not increase to cancel the parts count reduction. The biggest initial users of Co and Cr in an engine are discs, cases, and combustors; but blades and vanes become major users over the engine's lifetime as they are required in the largest quantities for spare parts (40).

Related to improvements in component efficiency are improvements in turbine airfoil cooling efficiency. In those engines which employ cooled airfoils (and some cannot because the airfoils are too small) this permits higher engine performance for the same airfoil surface temperature. Alternatively, it permits lower material temperatures and longer blade life with a given engine performance. However, turbine airfoil cooling technology is presently in a very advanced state, and only minor benefits will be derived from further advances.

3. **A new engine concept.** At some future time it is possible that a radically new type of engine will replace one or more of the aircraft gas turbine engine configurations. The new engine may either reduce or increase strategic material usage--in fact, new materials may become strategic because of it. Such a replacement engine does not exist today.

CONCLUSIONS

The aircraft gas turbine industry is in business to make money. It is therefore interested in improving the performance of engines while keeping the price of those engines acceptable to the market. This means that the industry seeks cost-effective increases in engine operating temperature and/or increases in engine life. This translates to a demand for continuously increasing temperature capability for hot-section materials. Today, these materials are the superalloys and they obtain much of their remarkable high temperature capability from several strategic materials, amongst which are cobalt and chromium. In the absence of a strategic materials shortage or a substantial price rise for these materials, the industry will not consider research and development expenditure on strategic material substitution unless such substitution promises performance and cost benefits.

Cobalt (Co) and chromium (Cr) are present over a range of composition levels in the majority of the superalloys used to fabricate aircraft gas turbine engine hot-section parts. Their presence affects some 16 material properties in these alloys; most of these effects are small, a few are moderate, and some are large, and nearly all are beneficial. The superalloys are complex metallurgical systems and there is no simple way of decreasing their Co and Cr content through alloy modification without suffering material property losses which are unacceptable to the industry.

This does not mean that alloy modification is not a possible or viable means to reduce Co and Cr usage in these alloys. Alloy modification does promise reductions but certainly not the elimination of Co and Cr. It is in fact one of many substitution options that together promise substantial Co and Cr savings; individually these vary in their potential for substitution and chance for actual use.

The various substitution options may be divided into four classes: alloy substitution; the use of coatings; the use of advanced structural materials; and substitution by design.

1. **Alloy Substitution.** Four possible approaches have been identified.
   a) Alloy replacement is the substitution of an existing or new alloy that contains less Co and/or Cr. In some cases this will result in

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lower engine performance or decreased component life. In others it can give both increased performance and life, but at increased cost.

b) Alloy modification. By this process we may hope to back Co and Cr out of our current alloys without performance loss. Laboratory tests with a few alloys indicate that it may be possible to decrease and perhaps eliminate Co from many alloys with impunity in the laboratory; however, some in industry claim that this is not a likely prospect in the production environment because of the process tolerance limitations of production heat-treatment equipment. The issue remains unresolved. It is impossible to remove Cr from these alloys without providing oxidation and corrosion protection by some other means such as a coating. At present, coating technology cannot provide a coating that has virtually perfect reliability. In the absence of such reliability, structural alloy Cr levels cannot be decreased much when a coating is used because of the risk of coating failure. Alloy modification therefore has limited potential for strategic material substitution. Further, the industry cannot afford the extreme expense of modifying even the alloys that are currently the major users of Co and Cr. Alloy modification will only be practiced during the development of new alloys as these are needed in the normal course of events for performance improvement.

c) Local property tailoring is the use of more than one material in a component, each lending one or more major material properties in specific areas of the component. This requires that the component be fabricated of smaller parts joined together, each made from one material. Joining technology for this application requires considerable development but the payoff in strategic materials savings can in some cases be large--e.g., dual-alloy discs.

d) Grading of metal powder used to make a component is a variation on local property tailoring. It takes advantage of powder metallurgy to continuously vary materials and properties. This technology appears to be technically feasible but is so far undeveloped.

2. **Coatings** may be used in two ways to reduce Co and Cr usage.

a) Greater reliance on coating protection against oxidation and

corrosion would allow Cr, and to a lesser extent Co, to be removed from the underlying material (substrate). Greater reliance on thermal barrier coatings may allow the use of substrate materials with lower temperature capability and reduced strategic material content. Until coating reliability becomes virtually absolute there is extreme reluctance amongst engine manufacturers and operators to pursue this course. Some small Co savings in substrate materials may accrue from the wider use of wear resistant coatings.

b) Thermal barrier coatings can be used to reduce substrate temperatures thereby extending component life, reducing component replacement rates, and saving Co and Cr over the life of the engine. This is a cost-effective technique for strategic materials savings. However, some major problems with thermal barrier coating integrity must be overcome before these coatings can be applied to turbine airfoils.

3. **Substitution of Structural Materials Under Development.** A range of new structural hot-section materials are in various states of research and development. These are:

- Rapid solidification processed superalloys
- Long range order systems (intermetallics)
- Directionally solidified eutectics
- Oxide dispersion strengthened (ODS) superalloys
- Fiber-reinforced superalloys
- Monolithic ceramics
- Ceramic-ceramic composites
- Carbon-carbon composites

Only rapid solidification processed superalloys and ODS superalloys are likely to see manned aircraft service in the next 5 years (ODS superalloys have already been introduced for turbine vanes). Ceramics and carbon-carbon composites are much further away—they may not be available for man-rated engine service until around the year 2010.

The Co and Cr content of the majority of these materials is negligible or zero. As a group they may be used for virtually all hot-section parts;

individually, most are limited in their application. They may be used in two ways:

a) individually to fabricate complete components

b) selectively throughout the component in local property tailoring--this may be performed in conjunction with current superalloys.

Only individual fabrication has attracted serious attention to date; the alternative would require significant advances in joining technology.

4. **Substitution by Design.** Three ways of saving Co and Cr by design have been identified.

a) Performance downgrading. The designer can extend component lives by designing them for use at temperatures and stresses somewhat below the maximum possible with the materials used. This will reduce component replacement rates. This strategy is in direct opposition to past and present design practice which generally requires that materials be pushed to their limit.

b) Increasing component efficiency can reduce the number and size of components for a given engine performance. This reduces strategic materials usage in the initial engine build and then throughout its life as components require replacement. A large reduction in the number of hot-section components is expected in the next generation of military engines because of improved component efficiency.

Improvements in turbine airfoil cooling efficiency can also lead to reduced component temperatures and life extension. In the future, such improvements are likely to be small because cooling efficiency is now approaching its limit.

c) New engine concepts to replace the various forms of the aircraft gas turbine as we know them might reduce strategic materials usage, or it might expand it. However, such engines do not exist today.

There are indeed numerous present and possible future technologies that may be used to substitute for Co and Cr in the aircraft gas turbine engine hot-section. Many of these technologies yet face massive materials research and development costs and, should they prove viable, large scale-up and qualification costs. Much of this cost will be borne by industry. Consequently, these technologies will never be introduced unless there is sufficient economic incentive. They will never be introduced for the sole purpose of reducing the consumption of strategic materials.

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Appendix

DUAL-ALLOY TURBINE WHEELS FOR SMALL GAS TURBINES

Garrett Turbine Engine Company--Phoenix, Arizona

Small gas turbines of 1000 or less shaft horsepower have widely utilized integral turbine stages in which the blades are a portion of an integral casting (a turbine wheel consisting of a disc plus blades) of a nickel-base superalloy. The small gas turbine industry uses integral turbines of both axial flow and radial flow designs. The prime motivation for the use of integrally cast turbine stages has been minimum cost. A secondary motivation in some designs is the inability to use separately inserted turbine blades due to excessively high stresses in the blade attachment region.

While integrally cast turbine wheels are attractive from a cost standpoint, they are unable to meet some long life requirements due to the inferior low-cycle-fatigue life of cast alloys when compared to forged or powder metal nickel-base superalloys. These latter alloys have better tensile and low-cycle-fatigue strengths than the cast alloys--but are inferior to the cast alloys in creep strength at high tempertures. An attractive solution to using the best properties of both types of superalloys is dual-alloy integral wheel construction, wherein a bladed ring of an appropriate high creep strength cast alloy is metallurgically bonded to a powder alloy with high tensile and low-cycle-fatigue strength. Three technical developments of the last decade have made dual-alloy integral wheels viable. These are: (1) the development of high temperature, high pressure autoclaves to permit bonding by hot-isostatic pressing (HIPping) of the two parts and (2) the development of high sensitivity ultrasonic inspection equipment to permit the inspection of the critical bond zone for defects, and (3) the development of high strength powder metal superalloys whose processing temperatures are compatible with hot-isostatic pressing temperatures (typically around 2200F).

Garrett performed research and development work in 1977 and 1978 that proved the technical viability of the HIP-bonded dual-alloy wheel concept. Both axial flow and radial inflow turbine wheels were fabricated and subjected to limited testing to prove the concept's feasibility.

The first opportunity to apply the technology to a new Garrett product arose in 1979, when full-scale development of the GTCP331 auxiliary power unit (APU) for the Boeing 757 and 767 began. Boeing imposed requirements for the cyclic life of the APU that were essentially the same as for the main propulsion engines. A preliminary design analysis rapidly indicated that an integrally cast third stage turbine wheel for this APU would not meet Boeing's life requirement. Two alternate designs were then considered: a bonded dual-alloy design utilizing a cast blade ring of IN713LC (a 0% cobalt-12.5% chromium alloy) and a powder metal hub of Astroloy (an 18.5% cobalt-15.0% chromium alloy) or a part 100% machined from a forging of Waspaloy (13.5% cobalt-19.5% chromium). Cost estimates indicated the dual-alloy part to be lower in cost than the machined Waspaloy part. Furthermore, the dual-alloy part could be volume produced with no additional capital investment, while a significant investment in new machinery would have been required for the machined Waspaloy part at the same production rate.

Since the dual-alloy wheel was very attractive economically, a major development program was launched to make it a reality. This program had the following significant elements:

- A major mechanical design effort to determine optimum bond location and determine critical defect size at the bond line.
- A major manufacturing process development effort.
- Development of a high reliability ultrasonic inspection process to detect critical size defects.
- Determination of critical material properties of the bond and of the powder Astroloy.
- Cyclic spin pit testing of bonded wheels to destruction.
- Testing of wheels in engines for times over 1000 hours and for over 6000 start cycles.

After successful completion of the development program, approval was secured from Boeing to use the dual-alloy wheel in production GTCP331 APUs.

The Garrett history on the GTCP331 dual-alloy turbine wheel plus related research and development work now makes it possible to utilize dual-alloy wheel technology in high temperature uncooled integral wheels. Experimental

dual-alloy wheels have been made using an equiaxed grain version of MAR-M247, a better high temperature alloy than IN713LC. This alloy has a 70F improvement in operating temperature over IN713LC. A recent development has been the casting of blade rings of the MAR-M247 alloy having directionally solidified (DS) turbine blades. The directional structure adds another 50F of temperature capability over the equiaxed grain form of the same alloy.

In addition, advanced foundry techniques can be used to produce blade rings having a single crystal structure. Depending on the alloy selected, this single crystal structure can add another 100F of temperature capability over the DS structure.

The first dual-alloy components made using a DS MAR-M247 structure in the turbine blades were radial flow turbine wheels for the AGT101 automobile gas turbine being developed by Garrett for DOE. These wheels (which have powder Astroloy hubs) will be used in an interim version of the engine where all the stationary hot-section parts of the engine are ceramic and the turbine wheel is the dual-alloy metal wheel. This engine will be capable of operation at a 2100F turbine inlet temperature, 500F higher than the original all metal engine with a turbine rotor machined from an Astroloy forging.

Although the hub material used to date by Garrett for both the axial and radial turbine wheel has been powder Astroloy, the dual-alloy concept can allow the selection of materials, within design restrictions, that are lower in critical elements such as Inconel 718 (0% cobalt). Thus, the dual-alloy wheel concept permits extreme design flexibility.

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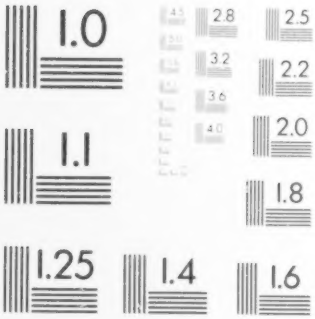

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